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ADVISORY GROUP FOR AEROSPACE RESEARCH & DEVELOPMENT

AGARD LECTURE SERIES No.168

Superplasticity

(Revised Version of Lecture Series 154)

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NORTH ATLANTIC TREATY ORGANIZATION

ADVISORY GROUP FOR AEROSPACE RESEARCH AND DEVELOPMENT

(ORGANISATION DU TRAITE DE L'ATLANTIQUE NORD)

AGARD Lecture Series No.168

SUPERPLASTICITY

(Revised Version of Lecture Series 154)

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THEME

Superplasticity has been transormed from a metallurgical curiosity to an important production process, particularly for low-to-medium production runs of components for the aerospace industry.

The whole spectrum of superplasticity was originally covered in Lecture Series 154 in the autumn of 1987, but such are the rapid advances in this technology, that the series will be re-presented, with the same speakers updating their lectures to give those attending the latest information on this most relevant technology and its impact upon the manufacture methods employed for components for aerospace applications.

This Lecture Series, sponsored by the AGARD Structures and Materials Panel, has been implemented by the Consultant and Exchange Programme.

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La superplasticité, qui ne fut à l'orgine qu'une curiosité en métallurgie, est aujourd'hui un procédé de fabrication important, qui trouve des applications dans l'industrie aérospatiale en particulier, pour des petites et moyennes séries de production de composants.

Le sujet de la superplasticité à été traité de façon très complète lors du Cycle de conférences No.154 organisé en automne 1987, mais il s'agit d'une technologie en plein essor, et les progrès réalisés dans ce domaine sont tels qu'il a été décidé de représenter ces Cycle de conférences. Ainsi, les conférenciers du Cycle de Conférences No 154 ont mis à jour leurs présentations de façon à y inclure les toutes dernières informations sur cette technologie intéressante et son impact sur les méthodes mises en oeuvre pour la fabrication de composants en vue d'applications aérospatiales.

Ce Cycle de Conférences est présenté dans le cadre du programme des consultants et des échanges, sous l'égide du Panel AGARD des Structures et Matériaux.

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Introduction

Certain alloys, when deformed in tension at particular temperatures and strain rates, show very high elongations. This phenomenon is termed superplasticity. Importantly, the flow stresses at which this phenomenon occurs are extremely low.

There are two types of superplasticity,

1. Isothermal superplasticity (ISP)
2. Cycling superplasticity (CSP)

and only the former will be dealt with in this paper.

For ISP, the alloy must possess an ultra-fine grain size which is relatively stable at $\geq 0.5T_m$, where T_m is the melting point of the lowest-melting constituent in the alloy, in absolute units, while for CSP the alloy must be capable of being cycled through a phase change.

The term 'superplasticity' is a translation from the Russian and became the universal term for the phenomenon in the 1960's, though it had been seen in deformation experiments as early as 1920.

Stable and unstable tensile flow

In ductile metals and alloys, tested at $<0.4\ T_m$, where plastic deformation occurs through dislocation glide and interaction, a typical load-extension curve is of the form shown in Figure 1. After load instability (P_{max}) the curve falls, due to diffuse necking, so-called because initiation is diffuse and propagation takes place over a relatively large area; in sheet metal the length of the neck is about equal to the specimen width. After the diffuse neck has developed to a certain stage - again in sheet metal - a local neck occurs. Here the test piece thins in a narrow band, at some angle to the applied tension and failure rapidly follows, as the volume of metal now deforming is small.

Local necking contributes virtually nothing to the total elongation of the specimen; for example, autobody steel sheet, showing 50% total elongation (50mm gauge-length) would divide into uniform elongation ~34%, diffuse elongation ~15%, necking contribution ~1%. If a stress (σ) strain (ε) curve is derived from the load-extension curve of Figure.1, as shown in Figure 2, it is seen that from zero strain to fracture strain hardening is occurring; the relationship between stress and strain is non-linear up to P_{max} and linear there from to fracture. The non-linear relationship can be quite well described by:

$$\sigma = k\varepsilon^n$$

k = a constant
n = strain hardening exponent

Load instability (dP = o) occurs when $\frac{d\sigma}{d\varepsilon} = \sigma$ and at this point $\varepsilon^* = n$, where ε^* is the limit of uniform true strain. In most metals $n \leq 0.5$ and this limits uniform elongation to ~60%.

The question now is what happens when the temperature is raised and thermally-activated processes are allowed to intrude? These processes involve time and so strain-rate must somehow be incorporated into the relationship.

We may write:

$$\sigma = k\varepsilon^n\dot{\varepsilon}^m$$

$\dot{\varepsilon}$ = strain-rate
m = strain-rate hardening exponent

Under strain-hardening conditions, $<0.4T_m$, m = 0 ± 0.03 and so the earlier relationship is still valid and if it is assumed that we are working at temperatures and strain rates such that strain hardening does not occur, i.e., the mobile dislocation density remains about constant then n = o and

$$\sigma = k\dot{\varepsilon}^m$$

Under these so-called hot working conditions, m~0.2. It should be further be noted that for m = 0.2, increasing the strain-rate by a factor of 10 will raise the flow stress by ~16%.

Values of m can be calculated from:

$$\underline{m} = \frac{\log (P_A/P_B)}{\log (V_2/V_1)}$$

These parameters are defined in Figure 3. It is clear from the foregoing that m is temperature-dependent. Figure 3 shows the change in slope at ~0.4 T_m for m, for a range of metals.

A load-extension curve for hot-working is shown in Figure 4. Compare this with Figure 1. What is clear is that the contribution of diffuse necking is now the major one to the total elongation.

In the case of pure aluminium, tested at 400°C and a strain rate of 1×10^{-3} sec^{-1}, the total elongation of 60% comprises 55% from diffuse necking.

Why is this? If:

$$\sigma = k\dot{\varepsilon}^m$$

and $\sigma = \frac{P}{A}$ P = load

A = cross sectional area

then $\frac{P}{A} = k\dot{\varepsilon}^m$ l = gauge length

as $\dot{\varepsilon} = \frac{1}{l}\cdot\frac{dl}{dt} = \frac{1}{A}\cdot\frac{dA}{dt}$

So, the shrinkage rate, $-\frac{dA}{dt}$

$$= \left(\frac{P}{k}\right)^{\frac{1}{m}} \cdot \frac{1}{A^{\frac{(1-m)}{m}}}$$

Thus the shrinkage rate is inversely proportional to the cross section - and highly sensitive to m. Thus, increasing m adds stability to the diffuse neck, and increases the diffuse elongation, as increasing n increases uniform elongation. The effect of m is illustrated in Figure 5, which shows that, as m approaches unity, the reduction rate at all cross-sections approaches a common level, and, at m = 1 $\frac{-dA}{dt}$ is independent of A, so, an irregular test piece would maintain its irregularities during deformation. There would thus be no local strain concentrations and large elongations would result; this behaviour is well known in hot polymers, glass and pitch.

Thus, superplastic alloys, some of which can exhibit total elongations >2000%, also have high m-values, typically ~0.6, which, in the correct temperature and strain rate régime, prodigiously extend the metal's capability for diffuse necking.

History

The first reference to a stress-strain-rate relationship seems to be that of Rosenhain et al in 1920 (1). He was studying the mechanical properties of a whole range of Zn/Al/Cu alloys and one of them (later identified as a eutectic) possessed unusual mechanical properties, exemplified by the following Table -

Effect of loading TIME on mechanical properties

Time of loading (secs)	Tensile strength tons in $^{-1}$	Total elongation %
3.5	62	9
12.5	32.3	9
72	27.0	42
92	23.9	22
2700	16.3	52

Rosenhain, who was very taken with the then-current so-called amorphous theory, interpreted these results as showing that heavily-worked metals contained a large amount of amorphous materials, which dominated their subsequent deformation behaviour and this view led him into fierce argument with other metallurgists of the time. This paper contains a photomicrograph (Fig.6) showing the extremely fine structure of this alloy and later, Jeffries & Archer (2) attributed this deformation behaviour to fine grain-size. Jenkins (3) observed large elongations in Pb-Sn at very low strain-rates and remarked that deformation could stop in one diffuse neck, whilst another would develop elsewhere, other papers of the period by Hargreaves and Hills (4) observed the time-dependence of Brinell hardness in certain alloys of this type.

1934 saw the now-famous paper by Pearson, who studied tin-lead and bismuth-tin eutectics, achieved a tensile elongation of 1950% and the well-known illustration, Fig.7. Padmanabhan and Davies (5) have pointed out that grain boundary sliding was shown in Pearson's photomicrographs (Fig.8); in fact he concluded that flow at grain boundaries was all-important.

Finally, in this era, Cook (6) showed that, changing to modern symbolism,

$$\sigma = \sigma_o + k\dot{\varepsilon}^{\frac{1}{2}}$$

could represent this deformation behaviour.

The next stage was a reawakening of interest in Zn-Al alloys in 1945 by the Russians. They showed particularly the importance of composition to large elongations (Fig.9) and came up with some strange theories of superplasticity (as now termed) which seemed unacceptable in the West. Another lull, until around 1960, when. again in Russia, Presnyakov and his team identified many new superplastic systems and put forward new ideas on the mechanisms of deformation - again all unacceptable!

The culmination was a review paper by Underwood (7) in 1962, which triggered off research activity which is stilll going on today, but first of all generated the now-classic paper of Backofen, Turner and Avery in 1964 (8), which was concerned with superplasticity in the Zn-Al eutectoid, and described the phenomenology in terms of $\sigma = k\dot{\varepsilon}^m$, and laid down the central role of m. It also showed (Fig.10) that these large, quasi-stable tensile elongations could produce remarkable (for metals) shapes in biaxial tension.

The question now being asked was - what was the metallurgical mechanism which was operating during superplasticity? The answer to this question had to satisfy the wealth of experimental results shown diagrammatically in (Fig.11), i.e., the relationships between:

stress and strain
stress and strain-rate
m and strain-rate
stress, strain-rate and temperature
m_{max} and temperature

grain-size and m

and also the metallographic evidence -

Very little grain growth (often found)
Curvature of phase boundaries (in eutectics/eutectoids)
Destruction of directionality
Grain boundary sliding
Grain rotation
Low dislocation density after deformation
Textural changes

The answer to this large question is addressed in another section of this paper.

Relationship of m to total elongation

If homogeneous superplastic deformation is assumed, and that material obeys

$$\sigma = k\dot{\varepsilon}^m$$

and that $P = \sigma A$

then $$\frac{dA}{dt} = -\left(\frac{P}{k}\right)^{\frac{1}{m}} \cdot \frac{1}{A^{\frac{(1-m)}{m}}}$$

This has been shown earlier. If this last equation is now integrated and certain other assumptions made, the total elongation for a rate-sensitive material can be found from the relationship obtained:-

$$\text{Total elongation} = e_o + 100\left(\left\{1 - \alpha^{\frac{1}{m}}\right\}^{-m} - 1\right)$$

If e_o is taken to be 40% (typical rate-insensitive figure), Figure 12 can be drawn and compared with the experimental results of Woodford (9). Depending on the width of the window, this theoretical relationship can be seen as satisfactory or not, but at least it emphasises the strong m - dependence of total elongation. Woodford's results show a wide scatter, particularly high m and low total elongation. This is, in the main, due to cavitation.

Cavitation

Most of the early work on superplasticity was carried out on zinc-based alloys and on zinc-aluminium eutectoid in particular. This particular alloy was very resistant to cavitation, if deformed under optimum temperature and strain rate conditions ($0.77T_m$ and 10^{-2} sec^{-1}), cavities can only be found in the fracture region after ~1000% tensile elongation.

Unfortunately, most of the alloys based on higher-melting-point metals, developed for commercial applications, are prone to cavitation, with the exception of Ti/6 Al/4V.

The initiation of cavities must result from the "incompatibility" of the phases present in the microstructure, the presence of hard particles, which do not deform with the body of the material, (cf. $CuAl_2$ in Al-rich solid solution) or, in a 50:50 by volume alloy, two phases of widely differing hardness (flow stress) at the superplastic temperature.

Grain or phase size and stability will also play a part. Taking a simple, rigid, two-dimensional, regular-grain-size model, with deformation occurring by grain sliding/rotation, a small grain size will, clearly, for a given elongation, result in a larger number of smaller cavities than a large grain size Figure 13. Thus, if the accommodation process (processes) for flow is diffusion (self-, grain-boundary-, pipe-) then the smaller the cavity the more easily it is dealt with.

In a number of superplastic alloys, grain growth occurs during deformation and so the cavitation problem will be accentuated by grain growth, due to time at temperature. Figures 14 and 15 show plots of hardness vs. temperature for the α and β phases in zinc-aluminium eutectoid and 60Cu/40Zn brass, these two alloys representing non-cavitation and cavitation behaviour respectively. At $\sim 0.76T_m$ for Zn-Al, the α/β hardness ratio is 1.2, while for Cu-Zn $\alpha/\beta \simeq 7$. It should be noted that, at $\sim 0.45T_m$ the hardness of the α and β phases in Cu-Zn are the same and so, if superplastic flow took place at this temperature, cavitation (from this cause) would not occur. Unfortunately, the required strain-rate would be so slow that massive grain growth would occur ($\alpha\beta$ brass is prone to easy grain-growth) and so significant elongations would not be achieved.

The general recipe is thus very small, stable grains/phases with similar mechanical properties. Exceptions to this are the superplastic ultra-high-carbon steels developed by Sherby and his co-workers (10), which exhibit microstructures of very-finely-dispersed cementite in ferrite.

The hardness of ferrite at 700°C is $\sim$16 Hv while that of cementite is $\sim$90Hv ($\sim$ 5.5x) but cavitation is not observed. This must be due to the extremely stable, fine phase-sizes achieved - 0.1 - 0.2 m - which can allow satisfactory accommodation, by short diffusion paths. Incidentally, work on certain alloy steels at Cranfield (D5 and H13) show superplasticity in D5 with a mainly ferrite/cementite microstructure) but no superplasticity in H13, a more highly-alloyed steel with chromium and vanadium carbides present. These latter do not soften at 700°C, having a hardness of $\sim$700 Hv at this temperature, and this leads to extensive early cavitation and rapid fracture.

Cavitation is, of course, deleterious to so-called service properties - strength
ductility
fracture toughness
impact strength
creep
fatigue
corrosion

all these are reduced by presence of cavities.

Figure 16 shows some SN curves for superplastically deformed Al/Cu/Zr showing clearly the effect of increasing superplastic strain.(increasing volume fraction of cavities) on life.

Fortunately, cavitation can be avoided or removed. If, instead of forming a sheet at ambient pressure, it is formed against a back-pressure of about half its flow stress, cavitation will be prevented. (Fig.17).

Alternatively, if cavitated parts are subsequently subjected to appropriate hot isostatic pressure (HIPping), see Figure 18, then cavities are closed and satisfactory material properties are usually obtained.

<u>Production and maintenance of ultrafine grains</u>

Producing a fine grain-size in a pure metal by, for instance, heavy cold ($<0.3T_m$) rolling is not too difficult, but maintaining this grain size at $\geq 0.5T_m$ is, for recrystallisation with grain growth will occur very rapidly and thus very rapid deformation would be necessary, which seems to run contrary to the rules of superplasticity. Floreen (11) conducted an experiment on pure nickel, however, in which a heavily cold-rolled sample was raised to about 800°C at 300°C per second and pulled on uniaxial tension at 2500 mm per minute. This yielded 225% elongation and an $\underline{m}$-value of 0.38. The heavily cold-worked microstructure recrystallised to a grain-size of 80μm in 10 seconds at 820°C. This phenomenon has been termed 'temporary superplasticity'.

Naziri and Pearce rolled commercial-purity zinc at room temperature ($0.43T_m$) (12) to a thickness reduction of 90% and transformed the original $\sim$120μm grain size to a stable 1-2 μm one; the stability was attributed to the impurities, notably lead, which appeared to lie in grain boundaries and pin them. Elongations of $\sim$200% at 10^{-4} sec^{-1} at $0.43T_m$ were achieved. Subsequently, a Zn/0.4% Al alloy, was similarly processed and gave $\underline{m} = 0.4 - 0.5$ and an elongation of $\sim$ 500% at 7×10^{-4} sec^{-1} at $0.43T_m$.

It is interesting to note that a zinc-tungsten alloy, essentially a dispersion of $\sim$1% of insoluble tungsten in zinc was <u>not</u> superplastic, even though the grain size of the zinc was small. This has been confirmed with other systems containing incompatible non-deforming particles.

An easy way, however, to produce a very fine phase-size is to cast an alloy of eutectic or eutectoid composition and then hot-work it (hot roll to 80% reduction) severely to break up and mix the two phases. This has been used in many instances and phase size stability is almost always adequately maintained at the superplastic temperature.

Zn-Al eutectoid is a special case, where soaking at 375°C for 4 hours, followed by quenching in iced water and removing after 8 minutes, triggers a solid-state transformation, which produces a fine (0.5 μm grains) α/β, 50/50-by-volume microstructure, highly superplastic at temperatures below 278°C (invariant) and with great grain stability. Static annealing of this alloy for 25 days at 250°C increases the grain size only to 4 μm, (NB, it is, of course, also, possible to produce fine-grain Zn-Al eutectoid by a thermomechanical processing route).

Thermomechanical processing can also be used to refine alloys with a high percentage of second-phase which are not eutectics/eutectoids. Cu60-Zn40 (α/β brass) and Ti/6Al/4V, both with ~ 50/50 by volume microstructures are two of these. However, the copper-zinc alloy grain coarsens very rapidly - 3 μm increasing to 10 μm in 30 minutes - and also cavitates severely. Thus it is difficult to produce very fine grains, impossible to retain them and large elongations are not possible due to cavitation, even if the other problems were resolvable.

On the other hand, Ti-Al-V highly superplastic at ~ 950°C does not seriously grain-coarsen nor cavitate and is the current superplastic mainstay of the aerospace industry.

The other most widely-used alloys in this industry are aluminium-based (13). The first was Al-6Cu-0.5Zr, specially developed for superplasticity, with the limitations of the earlier superplastic Al alloys in mind. One rule-of-thumb is that the successful addition of zirconium to aluminium refines the grains. Additionally, the addition of an alloying element such as copper, will lower the stacking-fault energy and promote recrystallisation, whilst raising the recrystallisation temperature. This commercial alloy is available only in sheet form, and supplied in a very hard temper, heavily cold rolled. Optical microscopy cannot reveal any microstructure in this original state.

After annealing for about 40 minutes at 450°C an equiaxed ~3 μm grain size is revealed and elongations of 500 -1200% are available at strain rates about an order of magnitude slower than Zn-Al eutectoid. Grain growth occurs steadily but not too rapidly during deformation, with the original fine grain-size being due to the presence of a very fine precipitate of $ZrAl_3$, which inhibits grain-boundary migration and also the presence of $CuAl_2$ precipitates. Other alloys of this type with similar compositions have been developed. The second group of aluminium alloys is the Al-Zn-Mg-Cu series. These are available commercial alloys, used by the aerospace industry for some years. Here by a specially developed thermomechanical process, (Fig.19) a grain size of ~10 μm can be achieved. This alloy is superplastic at 525°C at strain rates an order of magnitude lower again than Al-Cu-Zr. The fine grain size here also relies on pinning of grain boundaries.

Transformations can be used to produce fine structures in steel; tempering of martensitic structures can leads to fine dispersions of cementite. Marder (14) obtained $\underline{m}$ = 3.5 at 3.3 x 10^{-5} sec^{-1} in Fe -C at 700°C but only 98% elongation. Failure was due to cavitation at α-Fe_3C interfaces. This is typical of a lot of early work on steel superplasticity, until Sherby et al's (15) work on ultra-high-carbon steels. They produced a microstructure of spheroidized Fe_3C (0.1 - 0.5 μm) in ferrite grain matrix (0.5 - 1.5 μm) by a variety of thermal and thermomechanical treatments. The simplest is by oil quenching and then repeatedly (10-14 times) cycling the alloy through the A_1 temperature. This gave elongations of ~300% at 1.6 x 10^{-4} and 650°C ($\underline{m}$ ~0.45).

However, by austenitising, then rolling during cooling through $\gamma + Fe_3C$ range to break up the pro-eutectoid cementite as it emerges from the austenite just below A_1 and then rolling to spheroidise the pearlite, a superplastic UHC steel was produced which gave 750% elongation at $\dot{\varepsilon} = 1.6 \times 10^{-4}\ sec^{-1}$ at 650°C. No cavitation was seen and this must be due to the extreme fineness and stability of the structure and the plasticity of Fe_3C at 650°C.

Powder metallurgy technology can be used to produce superplastic properties. Nickel-based superalloy powders are made from the melt and then powder-forged directly into components. It is thought that the γ' phase may be responsible for grain refinement, but it may be due to minute traces of oxide on the original powder surface. This would be deleterious in tensile deformation, but may be effective as a means of restricting grain growth in forgings.

Mechanisms for superplasticity

What metallurgical deformation processes can operate at $\geq 0.5T_m$ and how would these be modified - if at all - by the fact that a superplastic alloy has a micrograin size?

Diffusion Creep

Both Nabarro-Herring and Coble creep relate stress, strain-rate and grain size, but it has been shown that the calculated strain-rate is too small, by orders of magnitude. In addition, they predict a linear relationship between stress and strain rate, which is rarely observed, and deformation by stress-directed vacancy diffusion will produce a grain shape-change which is not observed either. It would also maintain any existing crystallographic texture, which is not the case; in general, textures weaken during superplastic deformation.

Dislocation creep

Here there are also equations relating stress and strain-rate (no grain-size term) both in the bulk material and in grain boundaries. It is agreed that these relationships do not describe superplastic behaviour, but as Edington and Cutler (16) point out, dislocation climb in grain boundaries could be of assistance to grain boundary sliding (q.v.)

Diffusion-controlled Dislocation Glide

This is a special theory, proposed by Chaudhari, where superplasticity is controlled by the glide of jogged screw dislocations. He fits Backofen's results to this theory with the aid of certain parameters in his equations, but the experimental observations, which reveal a distinct lack of dislocations during and after deformation make the model difficult to accept. Also, any dislocation interaction would increase the flow stress (which is not observed) and in metal working at $\geq 0.5T_m$ where slip clearly occurs - hot rolling at high strain rates, for instance - the deformation is not superplastic.

Grain boundary sliding

Many workers have favoured GBS, or a variant of it, simply because GBS has been observed at the surface of superplastically-deforming alloys and because as grains become smaller, gliding and rotation become easier. Surface observations are not wholly satisfactory, however, for grains here have an extra degree of freedom and can move in and out of the sheet plane. Additionally, relationships between stress strain rate, grain size etc., predict m = 1, which is not observed.

In the case of GBS it is of course necessary to propose an additional process, - grain boundary migration, a diffusional process, recrystallisation - to maintain coherency, for in the case of most of the early work on Zn-Al eutectoid, no cavitation was observed, even after elongations of 1500%

A semi-philosophical point is - is grain boundary sliding a mechanism?
what is the mechanism of grain boundary sliding?

Due to the unsatisfactory nature of these single deformation mechanisms to describe superplastic flow, ore theories were developed in which combinations of diffusion, dislocations and GBS were introduced.

It is important to note that in the late 60's, early 70's when much of this theory was proposed, the experimental data came from eutectic and eutectoid alloys, whose deformation behaviour was inevitably complex. This was also the era of the 1 million volt electron microscope and with the possibility of in situ experimentation, Naziri et al attempted, in 1972, (17) to settle the question once, for all. By that time it was generally agreed that three mechanisms operate during superplasticity to a greater or lesser extent:

1. Grain boundary sliding
2. Diffusional creep
3. Dislocation creep/dynamic recovery*

*(This mechanism, plus dynamic recrystallisation, will be dealt with more fully later.)

4 μm-thick specimens of Zn-Al eutectoid with an 1 μm phase size were prepared and tested over a range of temperatures. Stills from the cine-film recording are shown in Figs. 20 and 21, taken at the indicated time intervals, enabling the microstructural changes that take place during the deformation to be followed. The aluminium rich and the zinc-rich grains have been labelled to indicate the relative grain movements. Some of the difficulties met with in these experiments were the lack of resolution and the continuous contrast changes which occur during straining, but luckily the dark and light shades of the two phases greatly facilitated their identification. In addition, the analysis of the process was complicated by the overlapping of grains in the thickness direction. Figure 22 shows the deformation behaviour of a particular group of grains. The important point to notice is the contraction of the horizontal boundary between the the two grains to give, finally, a vertical boundary.

The reverse situation is occurring in Figure 23 on four adjacent grains where the vertical boundary is contracting to give, finally, a horizontal boundary. The two grain-boundary configuration changes are shown schematically. Another feature to be noted is the rounding of the interphase boundaries, due to mass transport. This process must be occurring by the stress-directed diffusion of vacancies either through the lattice or along the boundaries of the aluminium-rich phase.

However, it is clear that grain movements relative to one another, leading to grain boundary sliding, are also taking place. The question now is whether diffusional creep is an accommodation process or the source of deformation. As other workers have demonstrated, when diffusional creep is the source of deformation, grain-boundary sliding is is inevitable, but diffusional creep can be treated as an accommodation process for steady-state grain boundary sliding. For boundaries which are not exactly planar (down to an atomic scale) and have bumps or serrations on them, steady-state sliding is only possible if matter is transported from one part of the boundary to another, by lattice or boundary diffusion.

The aluminium-rich phase appears inactive in this series of stills, but some dislocation activity has been observed in this phase, but to date the resolution is poor. This dislocation motion would also contribute to the diffusion of the zinc-rich phase, through dislocation cores.

This process is known as 'grain switching', after extrusion experiments carried out on a two-dimensional bubble model by Ashby and Verrall (18).

Their contention is shown in Fig.24; one event produces a true strain of 0.55, with less material movement over shorter distances than any other model, and the quantitative relationship between stress and strain-rate contains no adjustable constants.

It thus seems as if, in-situ electron microscopy, plus the Ashby & Verrall theory has at last 'solved' the question of superplastic mechanisms. Not so, Bricknell & Edington (19) questioned the validity of the in situ work, stating that irradiation from the electron beam could enhance diffusivity and so change the accommodation process, though Naziri et al maintained that, based on earlier in situ creep experiments,no radiation effects were evident. Bricknell & Edington also pointed out the grain-switching model needs to be extended to three dimensions. Here, I believe, the matter rests.

Dynamic Recovery/Recrystallisation

Superplasticity can, very crudely, be thought of as a sort of fast creep! Alternatively, it could be a special form of slow hot working. So far, by implication it has been thought of as the former. Now it will be considered from the latter standpoint.

Often, hot working is simply defined as recrystallisation 'keeping up' with deformation, but this definition is inadequate, for many variables, some obvious, others not quite so obvious, contribute - limits of strain-rate and temperature, m-value, shape of the stress-strain curve, stacking fault energy

Hot working clearly needs time for metallurgical restoration processes to happen, so strain-rate is all-important. If 'hot working' stress, strain curves are plotted, at some constant strain-rate, these fall into two types, exemplified in Fig.26 and labelled A and B. Both curves have three regions which correspond to -

1. The strain hardening region with an increasing dislocation density and the formation of the subgrain structure.

2. Oscillations, which indicate the transition between setting up and steady-state.

3. Steady-state.

Why the difference in the oscillations between materials? In A, representing aluminium, α-iron, zinc, metals with a high stacking-fault energy, cross-slip and climb are easy and so polygonisation can occur, such that strain hardening is offset. This is termed dynamic recovery.

In B, representing nickel, γ-iron, copper, dislocations are extended - low stacking-fault energy - cross-slip and climb are difficult. Some sub-grains form, but these are rapidly replaced by recrystallised grains. This is dynamic recrystallisation. In both cases, deformation at a constant flow stress for a constant temperature and strain rate - is achieved.

If material is deforming at a constant flow stress, this implies a constant dislocation density and at $0.5T_m$ grain boundaries, barriers to dislocation glide at $0.3T_m$, become sources and sinks for gliding dislocations.

In hot working, a coarse-grained (50 μm) metal at a strain-rate of, say, 2×10^2 sec^{-1}, a dislocation cell network (polygonisation) is formed; the sub-grain size is of the order of 1 - 2 μm. This enables dislocations to be created, to glide and to be annihilated at the rate required by the externally-imposed strain-rate to give microscopic deformation at a constant flow stress. If the strain-rate were decreased, then the sub-grain network would expand, as less dislocations would be required. In the limit, when the sub-grain size equalled the grain size we would be in the primary creep regime.

In hot working, a fine-grained metal (1-5 μm) at a strain rate of 2×10^2 sec^{-1}, no sub-grain boundary network needs to be generated, as the grain size is of the size of the notional sub-grains. Grain boundaries are adequate sources and sinks. If the strain-rate is now suddenly decreased, it is not possible for the grain size to change rapidly - grain growth is a relatively slow process. So an excess of grain-boundary volume exists and this facility for material transport is one aspect of superplasticity.

In superplastic alloys which do not have 50/50 microstructures, where fine grains are pinned by precipitates, dynamic recovery/recrystallisation are almost certainly the operating mechanisms.

In C.P. zinc and Zn-0.4% Al investigated by Naziri & Pearce (20) dynamic recovery was specified as the deformation process, for zinc has a high stacking-fault energy and the activation energy for the process was measured as 22 kcals/mole; Jonas gives 23 kcals/mole for hot-working of zinc. These values are close to those reported for zinc self-diffusion.

For dynamic recrystallisation, - low stacking-fault energy, dislocation climb difficult, nucleation followed by recrystallisation - activation energies are much higher. In the Al-6Cu-0.5Zr alloy where dynamic recrystallisation is the predominant mechanism an activation energy of is found (that for Al self diffusion).

Activation Energies

Activation energy measurements, however, as a guide to the mechanism of superplasticity, have long been a subject of controversy. There is an excellent discussion of the whole subject in ref. ( ).

Briefly, the activation energy at constant stress, Q_σ can be obtained by plotting $\log \dot{\varepsilon}_T$ against $1/_T$ or $\frac{1}{m} \log \sigma_T$ against $1/_T$

The activation energy at constant $\dot{\epsilon}$ can be obtained by plotting $\log \sigma_T$ against $1/_T$.

These must be unequal, and, with the interdependence of σ, $\dot{\epsilon}$, T, d and m, it becomes well-nigh impossible to conceive and execute an experiment in which there is only one variable in this area.

At best, activation energies can perhaps provide confirmation of an operating mechanism, but not definite proof of a particular mechanism on its own.

What is clear is that superplasticity is a phenomenon, linked to high m-values, and this value is produced in different alloys by the operation of different metallurgical processes - or their combinations -

self diffusion
grain boundary diffusion
stress-induced diffusion
solute diffusion
dynamic recovery
dynamic recrystallisation

(and also the metallurgical processes in Cycling Superplasticity - phase changes and thermal anistropy, not dealt with in this paper). Grain-boundary sliding, often proposed as a mechanism of superplastic flow, is another controversial area. Lifshitz (21) pointed out that during diffusional flow grain boundary sliding is inevitable; indeed the activation energy for sliding is generally that of grain boundary. Recently, there has been much speculation on the structure of grain boundaries and various grain-boundary-sliding models have emerged from this, all dependant upon the structures proposed in the models.

All that can be usefully said at present is that, under optimum superplastic conditions, all boundaries appear to behave similarly, and also that, until more evidence is available, grain boundary sliding is the result of diffusion parallel to the boundary curvature.

Manufacture

The paper which set superplasticity on the course it is still running on today, reference (8) contained a photograph (Fig.10) of a 'bubble' blown in Zn-Al eutectoid and it was from this beginning that sheet forming of superplastics grew. As m-value is the same in compression as in tension, it was not surprising that superplastic forging was being developed quite soon after sheet forming started.

Backofen, however was dissatisfied. In December 1969 he went on record (22) as saying -

"When is industry going to get off its rump and put some of this to work? When will people discover that this is no longer a theory, but a hard, practical, established concept that can make someone a lot of money?"

With hindsight we can answer this question - "In about fifteen years". (and it depends on what you mean by a lot of money).

Thermoforming

This is shown in various forms in Figures 26, & 27. Figure 26 shows a simple four-part operation. Forming into a female die should be used where possible, because of its simplicity. Figure 27 shows male forming where a die is forced against the sheet to produce the required shape, plus air pressure. The advantage here is that - if this is an important requirement, of course - inside dimensions are controlled, and the bottom of the part is approximately at the original metal thickness, because of friction. The control of thickness is important in thermoforming, and two methods adopted from the plastics industry are frequently applied - billowing and plug assisting. In billowing, increased surface area of the blank - ideally to that required by the finished part - is generated by blowing a bubble of approximately the required final surface area into an empty cavity and then, by reversing the pressure, 'laying' this material over the punch to produce a component of a reasonably uniform thickness.

Plug-assisted forming is shown in Fig. 28. With this method, virtually no thinning occurs of the sheet in contact with the plug; nearly all the metal flow is in the walls of the part. Blow moulding is coming into use as a forming technique as superplastic alloys are made available in tubular form.

Forging

As in conventional forging, a blank of exact weight (volume) is placed in the die cavity and formed as the two halves of the tool come together. Superplastics flow readily, so to avoid flash, good mating must be produced between the two parts of the tool and high clamping pressures used, often several times the forming pressure. Dimensions can be held to 0.025mm on the component, and 0.075mm over the parting line. For easy component ejection a 2° draft is recommended and the only limitation on component section is that the part must be robust enough to be handled at some elevated temperature during removal from the tools.

These comments were applied specifically to Zn-Al eutectoid, but should be relevant to any superplastic alloy (m 0.5) at optimum forming temperature and strain rate.

In these Zn-Al forging experiments, difficulty was experienced in removing the forged part from the dies and an ingenious solution was found. The dies were heated to about 300°C, above the invariant temperature (278°C) at which Zn-Al eutectoid is transformed to the non-superplastic state. Plastic deformation, however, occurred before the slug had time to transform, but by the time the dies were opened the piece was hotter but stronger! The same idea can be applied in sheet forming of this alloy by the use of an overheated mould (die) contact with which - when the sheet has attained the required form - destroys superplasticity.

At the other end of the temperature scale, superalloys can be superplastically forged accurately to shape. In this area it is interesting to think of the die materials which can be used for forging at temperatures such as 950°C. IN 100, a complex Ni-based alloy, has a strength of about 250-300 MPa at 950°C and this should be adequate for forging at optimum m, with a flow stress of 50 to 100 MPa.

When is it economic to operate this sort of process, which requires

Expensive dies
Special heating
Protective atmosphere
Low throughput

Today, conventional forgings often have buy-to-fly ratios of 10:1 and at £16 per kilogram this is so wasteful, that money is well-spent on a more sophisticated manufacturing process.

For instance -

Machined Part Weight - 35 kg
Conventional Forging Weight - 80 kg
Superplastic Forging Weight - 53 kg
27 kg saved, £432, per forging, in material alone

A variant of forging, known in the UK as 'hobbing' is a well-established die-sinking technique, whereby a die cavity is formed through the impression of a male punch (the master hob) into a block of die material. The hobbed cavity has a good surface finish (as good as the master hob) and many die sets could be made from one master. Unfortunately, strain-hardening of the die material during hobbing leads to high hobbing loads and even to distortion of the hob itself.

Certain popular die steels (24) can be appropriately heat-treated to induce superplasticity and now hobbing with very accurate replication can be carried out at low loads, using a hob manufactured from, for example, Nimonic 90 which is fifteen times stronger than a superplastic low-alloy steel (D5) under hobbing conditions. This process, if developed further seems to have some attractive features for the production of forging dies, for D5 steel is the most popular low-alloy steel for die manufacture in the hot forging industry.

At the low temperature end, Zn-Al eutectoid has been used for dies for the plastics injection moulding industry often for short or prototype runs, while for higher production the Al-Si eutectic with better wear-resistance has been employed. Note that Al-Si while no use in tension due to severe cavitation, is useful in compression, where the imposed stress system does not favour fracture.

Other Processes

Two other developments have emerged since the advent of superplasticity in aerospace - HIPping and Diffusion Bonding.

HIPping has been briefly dealt with in the section on Cavitation.

Diffusion bonding, as a method of joining during superplastic forming, emerged with the discovery of superplasticity in the two-phase Ti-Al6-V4 alloy. This alloy was specially developed for the aircraft industry with low-temperature strength and ductility properties in mind and it was subsequently discovered that, after working in the two-phase field, a fine, stable, $\sim 3 \mu m$, microstructure is produced, which at ~925°C has an equal proportion of the α and β phases. It shows excellent superplastic properties ($m = 0.8$ at 950°C and 1×10^{-4} sec^{-1}). Importantly, TiO_2 is soluble in Ti at 950°C, and so, if two pieces of Ti are pressed together in an oxygen-free environment, bonding takes place.

Thus the process described as - 'the first really new method of making things since the industrial revolution' - superplastic forming and diffusion bonding - SPFDB, was developed (23). The procedure is:

1. Preclean (mechanically or chemically)
2. Coat areas NOT to be diffusion bonded with a stop-off compound
3. Assemble in forming tools
4. Introduce argon to purge out oxygen and nitrogen
5. Heat up to 925°C
6. Introduce argon at 7 bar to start forming
7. When part is shaped increase pressure to 60 bar to ensure good diffusion bonding
8. Cool to room temperature
9. Acid pickle to remove any surface contamination

In Figure 29, the SPFDB concept using 4 sheets plus initial welding, is illustrated for the production of a high-performance missile wing.

Due to the reduced labour costs, a 40% unit-production, cost-saving is shown over a wing manufactured from an aluminium-alloy-machined core with a welded-up riveted-on, titanium alloy skin; this latter also weighed 50% more than the SPFDB one.

Unfortunately, Al_2O_3 is not soluble in Al and so the process of SPFDB cannot be directly applied to other systems. Much research, however, is being directed towards the joining of superplastic aluminium-alloy-sheet during forming; practical developments from this research are eagerly awaited.

Service Properties

The usual in-service mechanical properties and the service conditions to which engineering components are subjected are:-

Properties - yield strength, tensile strength, ductility, impact strength, fracture toughness

Conditions - creep, fatigue, corrosion

Yield strength, Tensile Strength

The strength of a polycrystalline metal varies with grain size:

$$\sigma \propto d^a$$

For slip deformation, $\leq 0.3T_m$, $a = -\frac{1}{2}$, whether stress is measured as yield strength, tensile strength or hardness: the well-known Hall-Petch relationship states:

$$\sigma_y = \sigma_i + k_y d^{-\frac{1}{2}}$$

and in ferritic steels, k_y = 15 - 20 MPa. This amounts to, roughly, 70 MPa increase in yield-strength for every 10x reduction in d. Thus an alloy with a 1 μm grain size will, have good strength at $0.3T_m$.

Cavitation, (Fig.30) during superplastic deformation will also affect room-temperature,post-forming properties. Figure 31 shows the effect of superplastic strain on certain engineering properties of 7475 Al alloy in the T6 condition.(25).

Ductility

Ductility, here measured by total elongation on a 12mm gauge-length tensile test, again on 7475 T6, also falls with increasing cavitation - rather more significantly than do strength properties.

Impact strength, Fracture toughness

In a Charpy impact test, the presence of cavities will clearly lower the strength, though I could not find any published results. In the case of a 50/50 microstructure such as Zn-Al eutectoid, the zinc-rich phase is CPH and the Al-rich phase FCC. Thus as the temperature is lowered the CPH phase will show a ductile/brittle transition, while the FCC phase will not. It is interesting to note that CPH dominates and that the alloy exhibits a D/B transition.

Fracture toughness experiments have been carried out on formed, cavitated Al/Cu/Zr alloys in various conditions of heat treatment. The fatigue-crack growth rate increased with increasing cavitation - and with increasing strength attained by solution-treatment and ageing. This latter is thought to be due to a reduction in fracture toughness (26).

Creep

Clearly, if the superplastic forming temperature is close for the operating temperature, it is desirable for superplasticity to be eliminated after forming and before use. In the Zn-based alloys, room temperature is about $0.43T_m$ - too high for stable. post-forming behaviour. This was discovered when a motor-car body was manufactured from Zn-Al eutectoid sheet and stability was absent! The problem was alleviated in an interesting way; 1% copper was added to the alloy (27) which had no effect on m and superplasticity at 250°C, but reduced the steady-state creep rate at 20°C by a factor of 140.

In the superplastic superalloys, elaborate heat-treatments are often used to promote grain coarsening.

Fatigue

If an alloy possesses a strain-rate-sensitivity of the flow stress, and if fatigue life correlates with yield stress, then the life of a superplastic alloy should increase with increasing frequency. This has been shown to be so (28) in Zn-Al

On the other hand, superplasticity with cavitation - increasing with increasing superplastic strain - will have an adverse effect on life (Fig.32)

Corrosion

In 50/50 alloys, the corrosion behaviour can be controlled by either phase. In the Zn-Al eutectoid, atmospheric corrosion is slow, due to the formation of an air-formed film from the α-phase, making it behave similarly to an aluminium alloy.

If, however, a piece is joined (electrically) to steel, under wet conditions, it will dissolve rapidly away, as the β-phase is anodic to the steel, which it now protects sacrificially.

The Al/6Cu/0.5 Zr alloys possess poor corrosion resistance, as far as the whole spectrum of aluminium alloys is concerned and so cladding is applied for the manufacture of parts to be used in malign environments. It must be ensured, however, that the clad layer remains coherent, otherwise local attack and subsequent perforation may ensue.

Conclusion

In this overview I hope that I have interested you - and informed you - in some aspects of the huge subject of superplasticity. There may have been a slight pre-occupation with alloys not envisaged for aerospace applications, but nevertheless, they demonstrate principles which have to be borne in mind in all engineering applications.

I believe in the future that new superplastic alloys will continue to be developed and existing ones refined and applied for a wide range of engineering products.

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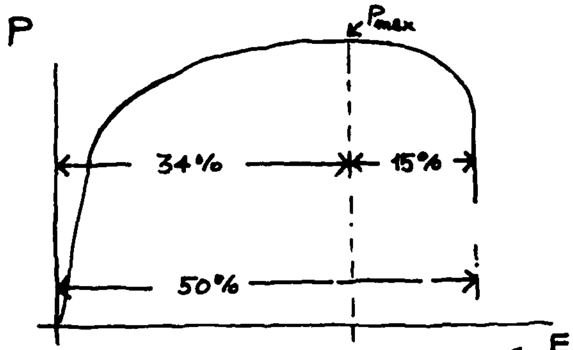
FIGURE 1 Typical load-extension curve for a ductile metal at $0.3T_m$



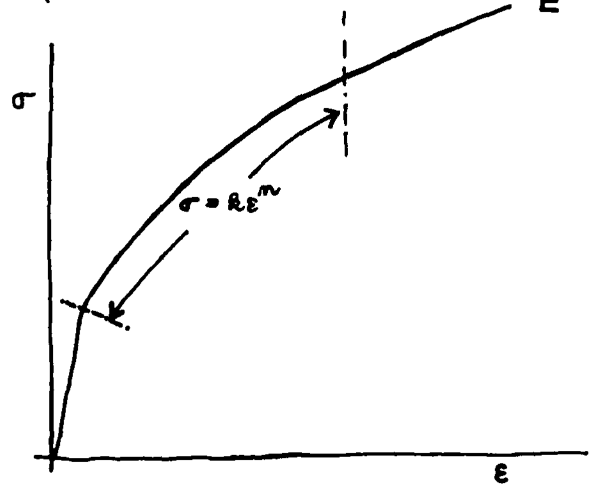


FIGURE 2 True stress-strain curve derived from FIG.1

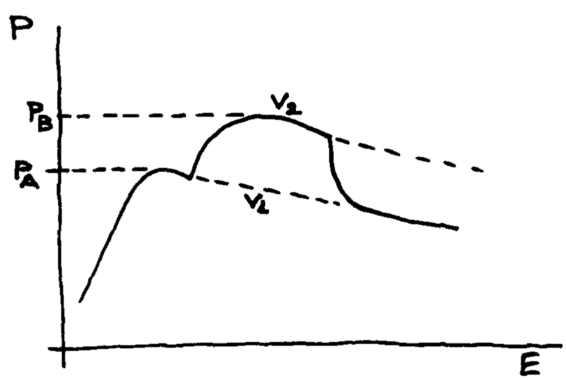


FIGURE 3 Determination of m-value

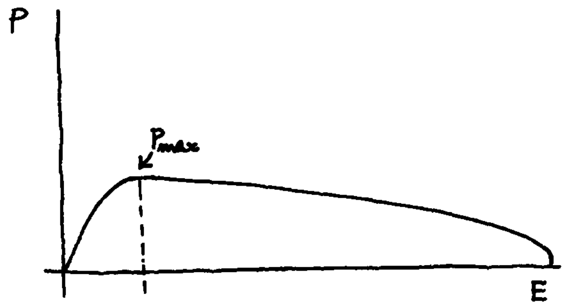


FIGURE 4 Load-extension curve for $>0.5T_m$

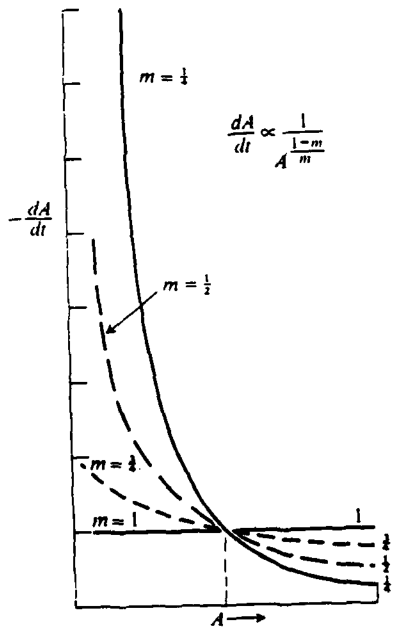


FIGURE 5 Effect of m on specimen shrinkage

FIGURE 6 Microstructure (x 650) of Rosenhaim et al's Al/Cu/Zn alloy showing SRS

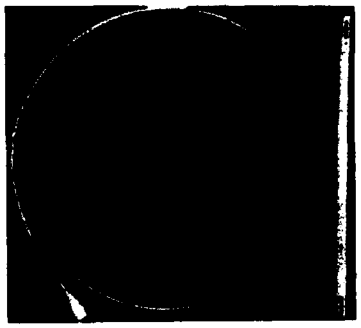


FIGURE 7 Pearson's tensile test

FIGURE 8 GBS indicated on one of Pearson's photomicrographs

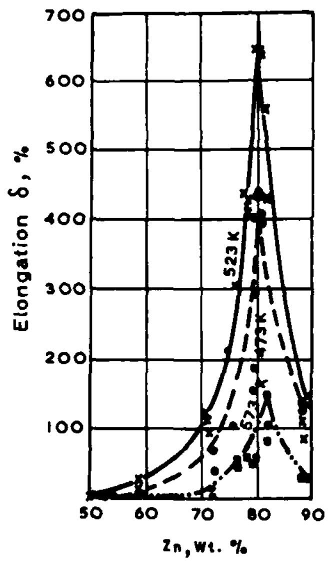


FIGURE 9 Effect of alloy composition upon superplastic elongation

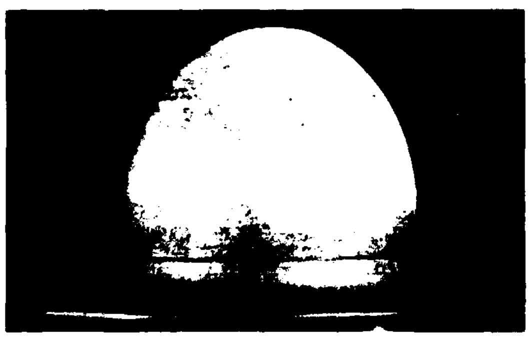


FIGURE 10 Original 'bubble' blown by Backofen et al

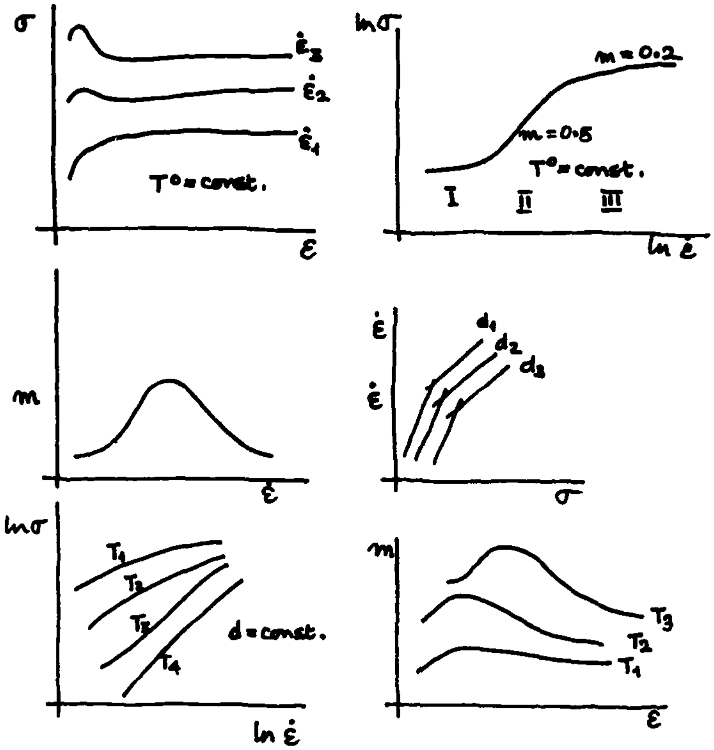


FIGURE 11 Effect of various parameters upon superplastic behaviour

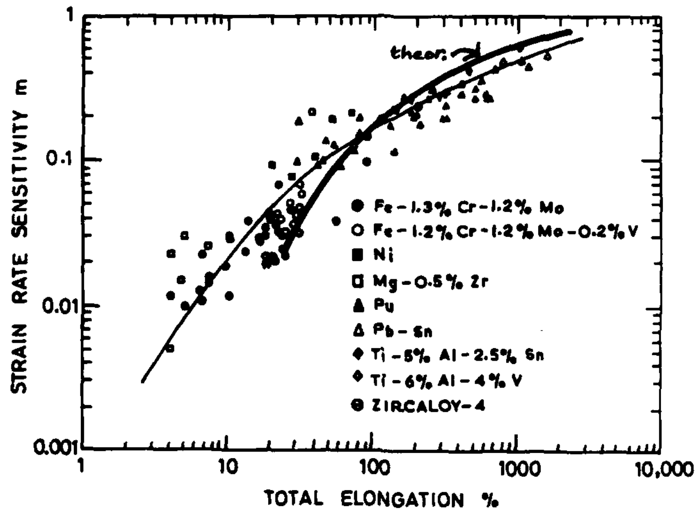


FIGURE 12 m versus elongation, both theoretically and experimentally

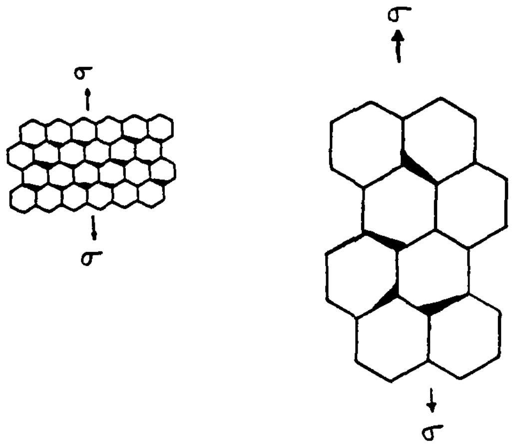


FIGURE 13 Effect of grain size on GBS

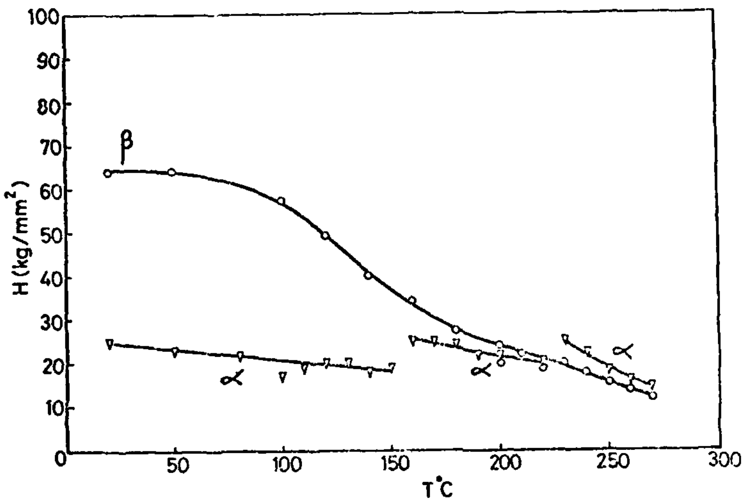


FIGURE 14 Hardness versus temperature for the α and β phases in Zn78/Al22 superplastic alloy

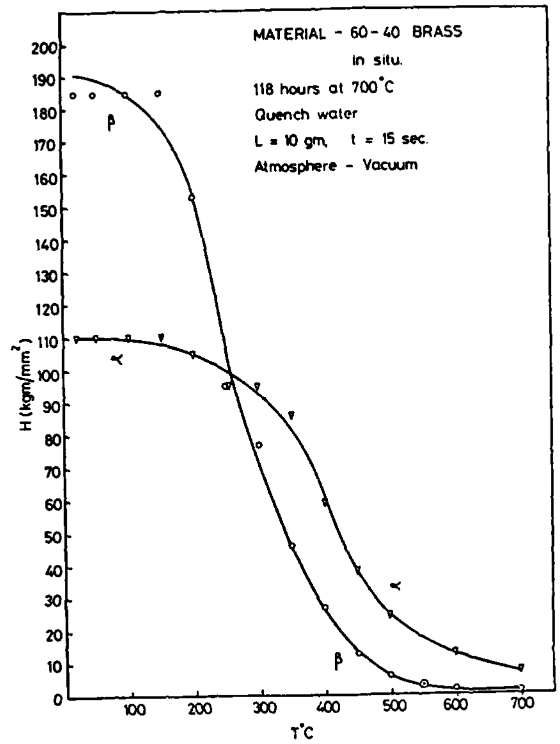


FIGURE 15 Hardness versus temperature for the α and β phases in Cu60/Zn40 superplastic alloy

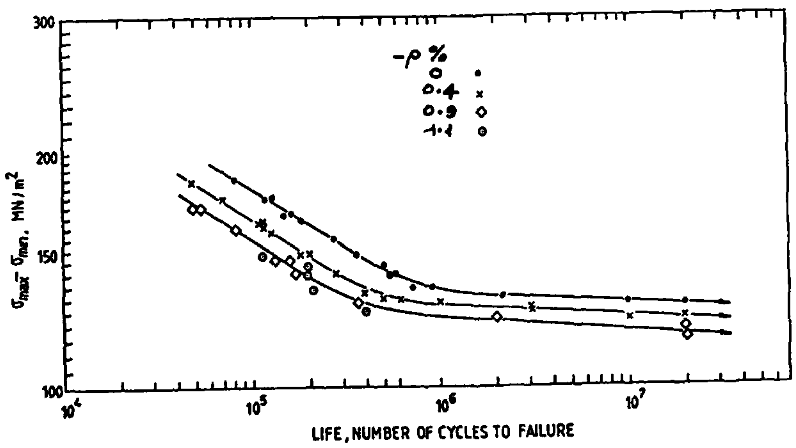


FIGURE 16 SN curves for Al/Cu/Zr superplastic alloy, showing the effect of cavitation

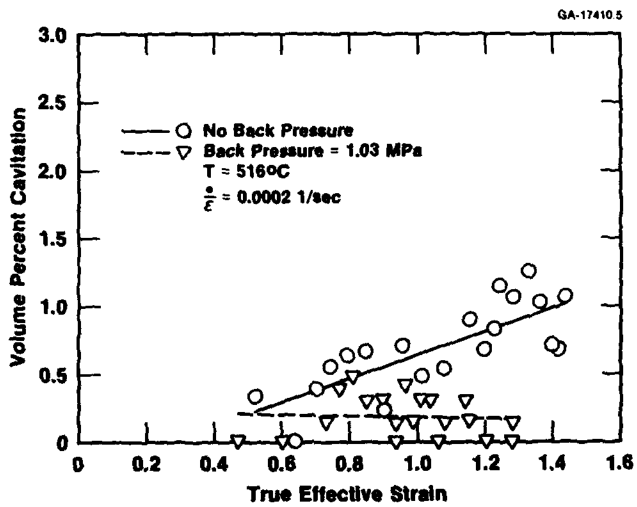


FIGURE 17 The effect of forming back-pressure on cavitation, with increasing strain

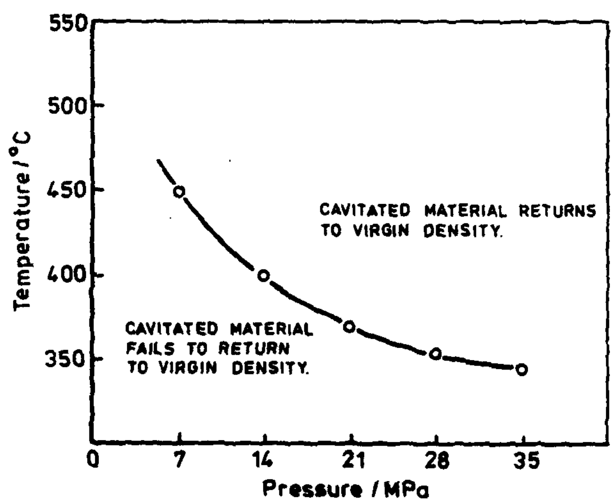


FIGURE 18 A temperature/pressure cavitation limit diagram

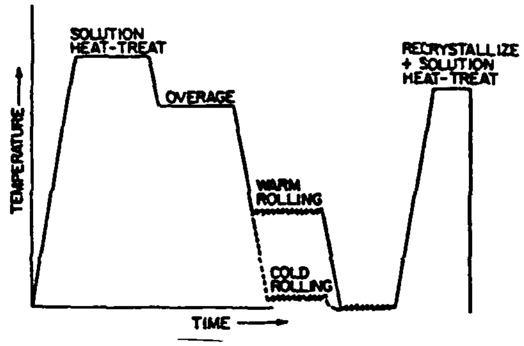


FIGURE 19 A production recipe for fine-grained Al 7475 alloy

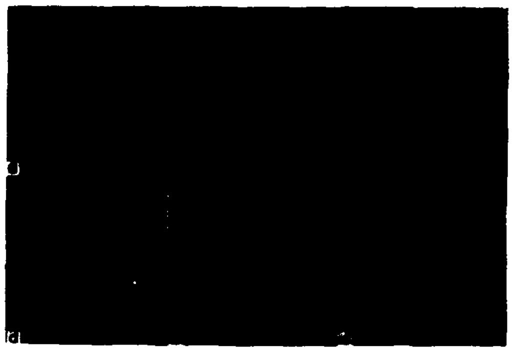


FIGURE 20 In-situ deformation in a HVEM of Zn/Al eutectoid

FIGURE 21 In-situ deformation in a HVEM of Zn/Al eutectoid

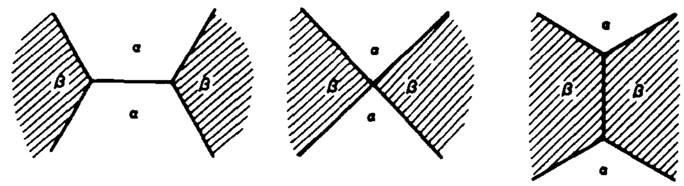


FIGURE 22 Diagrammatic representation of FIG. 20

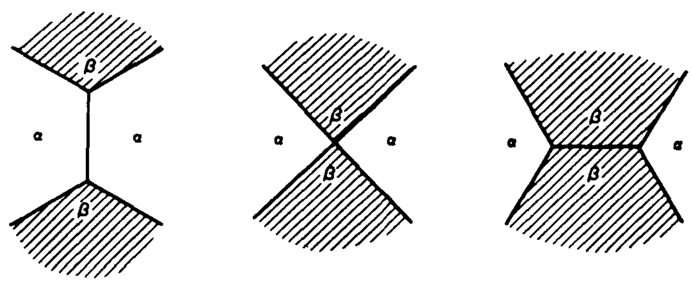


FIGURE 23 Diagrammatic representation of FIG. 21

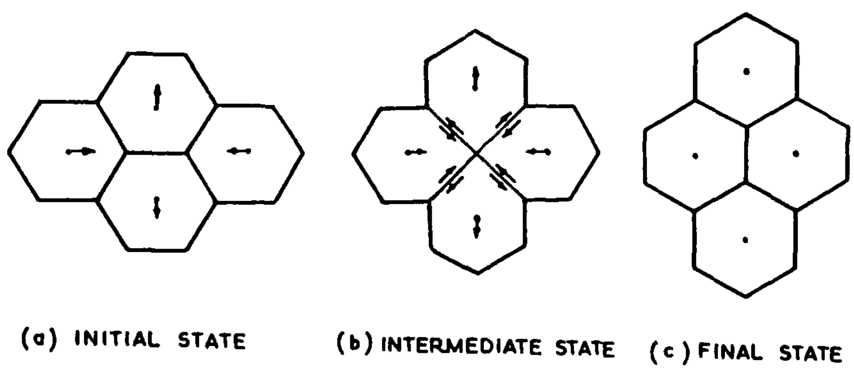


FIGURE 24 Ashby & Verrall theory

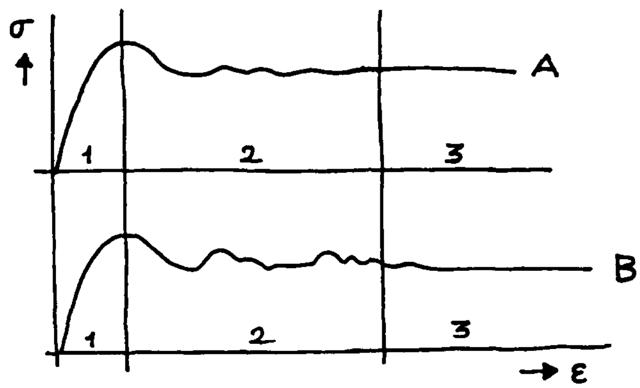


FIGURE 25 'Hot'-working stress-strain curves

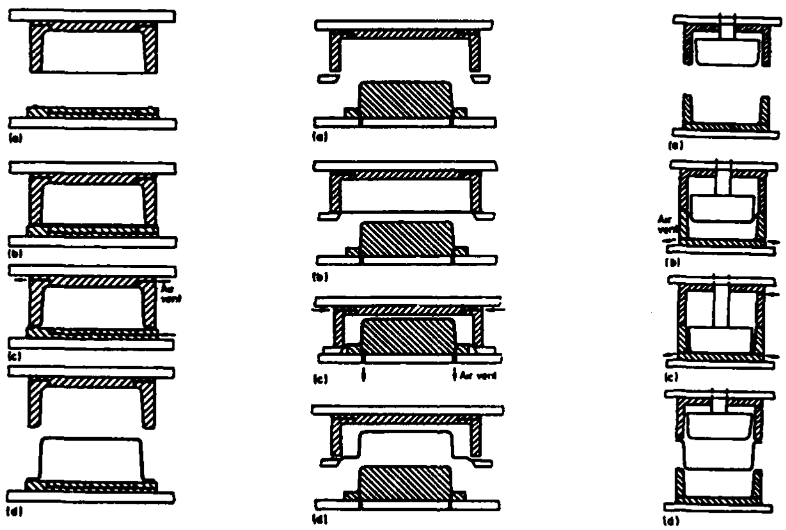


FIGURE 26, 27 and 28 Various superplastic-sheet forming methods

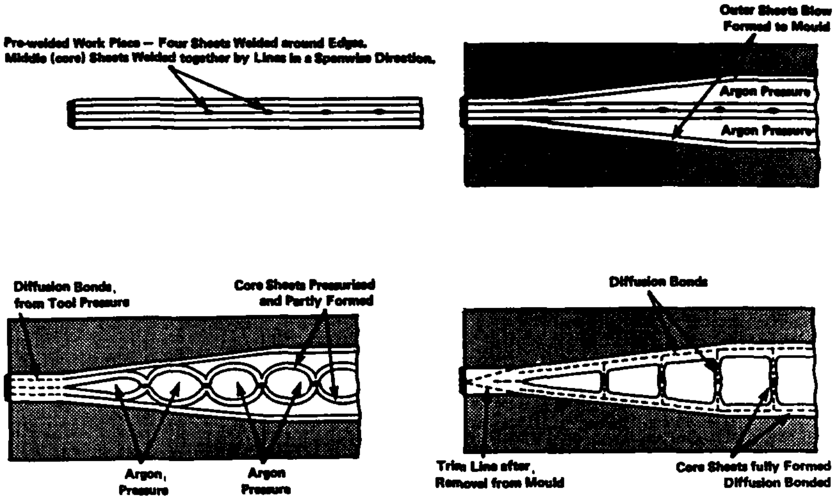


FIGURE 29 Advanced manufacturing, involving SPFDB

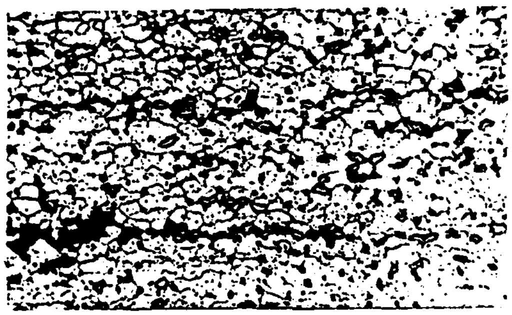


FIGURE 30 Cavitation in superplastically-deformed Al/Cu/Zr alloy

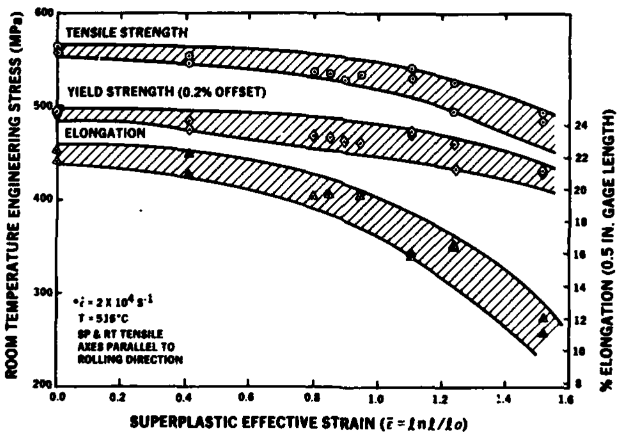
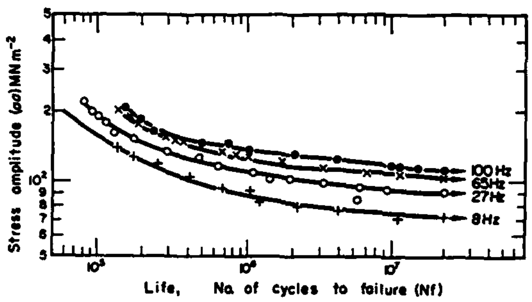


FIGURE 31 Effect of superplastic strain upon various room-temperature mechanical properties



Stress amplitude vs life curves for fine-grained SPZ at room temperature.

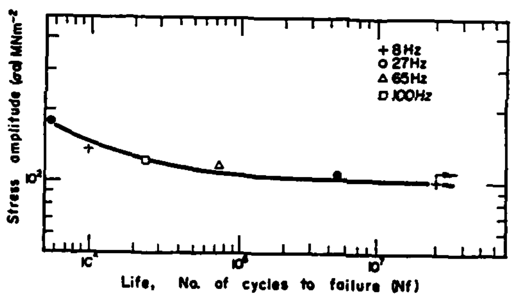


FIGURE 32 Effect of strain-rate-sensitivity upon fatigue life in Zn/Al eutectoid

SUPERPLASTIC SHEET FORMING
NATO/AGARD Lecture Series on Superplasticity

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Abstract

The exceptional ductility of superplastic alloys can be utilized in the shaping and forming of parts, components, and structures which could not be easily or economically produced by materials of more limited ductility. A number of methods for forming these materials have been studied on a laboratory scale, and several of these are being utilized to produce full scale parts on a production basis, with benefits being achieved in cost as well as in design efficiency. The forming of superplastic alloys, however, involves consideration of a number of factors which are interactive and lead to a relatively complex process, especially if maximum capability and minimum cost are to be achieved. These considerations include the superplastic properties of the alloy, effect of temperature, effect of strain rate, microstructural changes during forming and their effect on superplastic properties, effect of die configuration on forming capability, and forming parameters.

The forming methods which have been demonstrated for superplastic alloys include: blow forming, vacuum forming, thermo-forming, die-less drawing, deep drawing, forging, and superplastic forming with combined diffusion bonding (SPF/DB). With the exception of the forging, these processes utilize the high elongation capability and related resistance to localized necking. Therefore, the thinning gradients which can develop during forming are of primary concern, and relate directly to the material characteristics and the mechanics of stretch forming into a given die configuration. The technology of SPF has been found to benefit from modelling of the process, the results of which can guide the selection of pressurization parameters as well as predict the thinning characteristics and tendency to rupture. These concepts are reviewed in this paper. Many alloys tend to cavitate, or form internal voids. During SPF processing and mechanical design properties can suffer if this this cavitation is severe. Concepts for minimizing or eliminating cavitation are discussed.

Introduction

Superplasticity is a term which is used to indicate the exceptional ductility that certain metals can exhibit when deformed under proper conditions. The term is most often related to the ductile tensile behavior of the material; however, superplastic deformation has the characteristic of easy deformation under low pressures, and compression deformation characteristics are also described as superplastic. The tensile ductility of superplastic metals is typically in the range of 200 to 1000% elongation, but ductilities in excess of 5000% have been reported (1). Elongations of this magnitude are one to two orders greater than those observed for conventional metals and alloys, and are more characteristic of plastics than metals.

Since the capabilities and limitations of sheet metal fabrication are most often determined by the tensile ductility limits, it is clear that there is are significant advantages potentially available for forming such materials, if the high ductility characteristics observed in the tensile test can be utilized in production forming processes. This, of course, is being done today and the number of applications of parts formed by these methods is increasing each year. This paper will address many of the processes and related considerations important in the forming of superplastic sheet metal parts.

Requirements for Superplasticity

Before discussing the details of the superplastic forming processes, it is necessary to review the more important aspects of superplastic material behavior, since some of the specific forming parameters are determined by this behavior.

There are several different types of superplasticity in terms of the microstructural mechanisms and deformation conditions, including the following (2,3): a) micrograin superplasticity, b) transformation superplasticity, and c) internal stress superplasticity. At this time, only the micrograin superplasticity is of importance in fabrication of parts, and the discussion will be limited to this type. For the micrograin superplasticity, the high ductilities are observed only under certain conditions, and the basic requirements for this type of superplasticity are: 1) very fine grain size material (of the order of 10 microns), 2) relatively high temperature (greater than about half the absolute melting point), and 3) a controlled strain rate, usually in the range 0.0001 to 0.01 per second. Because of these requirements, only a limited number of commercial alloys are superplastic, and these materials are formed using methods and conditions that are different than those used for conventional metals.

Characteristics of Superplastic Metals

For a superplastic metal tensile tested under proper conditions of temperature, the observed ductility is seen to vary substantially with strain rate, as shown in Figure 1 for a Zn-Al eutectoid alloy (4). As shown, there is a maximum in ductility at a specific strain rate, with significant losses in ductility as the strain rate is increased or decreased relative to this maximum. It is now well known that the primary factor related to this behavior is the rate of change of flow stress with strain rate, usually measured and reported as m, the strain rate sensitivity exponent, where

$$m = \partial ln\sigma / \partial ln\dot{\epsilon}$$

and σ is the flow stress and $\dot{\epsilon}$ is the strain rate.

The characteristic flow properties for a superplastic metal are exemplified in Figure 2 for a Ti-6Al-4V alloy tested at 927°C. The m for these data is the differential of the above curve as indicated above, and a plot of m vs log strain rate is shown in Figure 3. The very strong strain rate sensitivity of flow stress as shown in Figures 2 and 3 is typical of superplastic metals and there is a good relationship between the m value and superplastic ductility, a relationship that was shown most clearly by Woodford (5) in a graph of data for a large number of alloys as shown in Figure 4 where the m value is graphed as a function of elongation. While the total elongation can also be affected by fracture, the strain-rate sensitivity is a first order effect. The influence of the m on the ductility is understood through mechanics to be due to the stabilizing effect of the strain rate sensitivity of flow stress on the diffuse necking process (6-8). The superplastic deformation is also strongly affected by temperature, and an illustration of typical behavior is shown in Figure 5 (9) where the ductility of a titanium alloy is graphed as a function of temperature. As can be seen, the elongation rises and falls rapidly over a relatively short temperature range, and outside the limits of this temperature span the ductility is quite modest and within the range of conventional material behavior.

The above characteristics of a superplastic alloy indicate, therefore, that unusual forming capability should be possible with superplastic alloys, but that control of the forming process parameters is important in order to obtain the full potential of this class of material. Such process controls are more demanding than corresponding requirements for conventional forming processes, and superplastic forming of sheet metals is a technology which is new and different from the conventional processes. However, when properly conducted, SPF offers advantages over other fabrication methods for a number of applications.

Superplastic Alloys

Because of the stable grain size requirement for a superplastic metal, not all commercially available alloys are superplastic. In fact, very few such alloys are superplastic. Many materials have been produced under laboratory or pilot-plant processing (3), but very few of these have been produced commercially. Nonetheless, there are some alloys which can be obtained, or which may be expected to be available in the future, that can be mentioned. It is expected that, as the SPF technology develops, there will be additional alloys produced specifically for this process.

A summary of several superplastic alloys is presented in the Table 1, along with some of their characteristics. Particularly noteworthy are the Ti-6Al-4V, 7475 Al, and the Supral alloys which are quite superplastic and are commercially available. The Ti alloys have been found to be superplastic as conventionally produced, and there has not been a need to develop alloy modifications nor special mill processing methods to make them superplastic. However, that has not been the case with the Al alloys, and either special processing (10) or alloy development (11) has been necessary to produce superplastic materials. The Zn-22Al alloy is one which has been the focus of substantial research because it can be readily processed into the superplastic condition, and this alloy has also been made available commercially by several different suppliers.

Important considerations in the selection and use of a superplastic alloy are the total elongation capability, the stability of the superplastic microstructure at high temperature, the latitude of the temperature and strain rate range over which superplasticity is observed, and the rate of development of cavitation during superplastic deformation. All of these factors can change from lot to lot of the material, and it generally advisable to check each lot for the superplastic properties as well as design properties.

SPF Processes

A number of methods and techniques have been reported (3,12) for forming superplastic materials, each of which has a unique capability and develops a unique set of forming characteristics. The following are forming methods which have been used with superplastic alloys:

Blow Forming
Vacuum Forming
Thermo-forming
Deep Drawing
SPF/DB

Forging
Extrusion
Dieless Drawing

Only those processes that relate to sheet metal forming are discussed in this paper.

Blow Forming and Vacuum Forming

Blow forming and vacuum forming are basically the same process (sometimes called stretch forming) in that a gas pressure differential is imposed on the superplastic diaphram causing the material to form into the die configuration (3,12-14). In vacuum forming, the applied pressure is limited to atmospheric pressure (ie., 15 psi) and the forming rate and capability are therefore limited. With blow forming, additional pressure is applied from a gas pressure reservoir, and the only limitations are therefore related to pressure rating of the system and the pressure of the gas source. Typically, a maximum pressure of 100 psi to 500 psi is employed in this process.

The blow forming method is illustrated in Figure 6 where a cross-section of the dies and forming diaphram are shown. In this process, the dies and sheet material are normally maintained at the forming temperature, and the gas pressure is imposed over the sheet causing the sheet to form into the lower die, and the gas within the lower die chamber is simply vented to atmosphere. The lower die chamber may also be held under vacuum, or a "back pressure" may be imposed to suppress cavitation if necessary. The concept of the use of "back pressure" to control or prevent cavitation is discussed in a subsequent section of this paper.

The rate of pressurization is normally established such that the induced strain rates in the forming sheet are maintained in the superplastic range, a rate which is either determined by trial-and-error, or by application of analytical modeling methods (15-18). This pressure is generally applied slowly rather than abruptly to prevent too rapid a strain rate and consequent rupturing of the part.

The perifery of the sheet is held in a fixed position and does not draw-in as would be the case in typical deep-drawing processes. It is common to use a raised land machined into the tooling around the perifery as shown in Figure 7 to secure the sheet from slipping and draw-in, and to form an air-tight seal to prevent leakage of the forming gas. The sheet alloy therefore stretches into the die cavity and all of the material to form the part comes from the sheet over-laying the die cavity. This results in considerable thinning of the sheet for complex and deep-drawn parts, and can also result in significant gradients in thickness in the finished part.

This process is being used increasingly to fabricate structural and ornamental parts of titanium, aluminum, and other metals. An example of the process applied to the forming of a titanium aircraft nacelle frame (19) is illustrated in figure 8. In this case, the forming is conducted at about 1650F, and inert gas (argon) is used on both sides of the sheet in order to minimize oxidation and related detrimental surface degradation due to the reactivity of titanium. The use of such protective gases is not usually necessary for aluminum alloys.

Large complex parts are readily formed by this method, and it has the advantage of no moving die components (i.e., no double acting mechanisms) and does not require mated die components. Multiple parts can be formed in a single process cycle thereby permitting an increase in the production rate for some parts.

Thermo-forming

Thermo-forming methods have been adopted from the plastics technology for the forming of superplastic metals, and may employ a moving or adjustable die member in conjunction with gas pressure or vacuum (13,14,20). An illustration of techniques employing this process are shown in Figure 9. In the first case, an undersized male die punch is used to initially stretch form the superplastic sheet followed by application of gas pressure to force the sheet material against the configurational die to complete the shaping operation.

Another method also illustrated in Figure 10 employs a movable die member which aids in pre-stretching the sheet material before gas pressure is applied. In this case the gas pressure is applied from the same side of the sheet as the moving die. These two techniques provide ways of producing different shapes of parts, and can be used effectively to control the thinning characteristics of the finished part.

Deep Drawing

While deep drawing studies have been conducted with superplastic metals, this process does not appear to offer many significant advantages in the forming of superplastic materials. Deep drawing depends on strain hardening to achieve the required formability and prevent thinning and rupture during forming. Superplastic materials do not strain harden to any great extent, but depent on the high strain-rate hardening for their forming characteristics, and this property seems to offer little aid to deep drawing.

The difficulty is that, in order to draw in the flange, the material in contact with the punch nose as well as that in the side wall must work harden to carry the increasing stresses required to draw-in increasing amounts of the flange. At superplastic temperatures, no significant work hardening occurs, and the punch typically pierces the blank, or the blank fails in the cup walls if the frictional constraint between the punch and the blank is high. However, Oshita and Takei (21), in studies on the Zn-Al alloy, were able to develop a maximum draw ratio under optimized conditions of 2:1.

A technique which tends to improve the drawability of superplastic alloys is that reported by Hawkins and Belk (22). This method, illustrated in Figure 11, utilizes a punch cooled to a temperature below that of the forming blank whereas the hold-down tooling is maintained at the forming temperature. In their studies, they demonstrated that this differential temperature technique permitted an increase in the limiting draw ratio (LDR) from less than 2.4:1 for isothermal conditions to more than 3.75:1 for the differential temperature method. The thinning characteristics for this process are also shown in the Figure 11. Slight thinning can be seen to occur over the (cold) punch nose, and substantial thinning is seen in the material adjacent to the punch, the extent of which depends depends on the blank hold-down load, but increases with increasing blank diameter (draw ratio) and decreasing punch speed. Since ironing was not utilized in these forming tests, thickness increases were observed at the greater distances from the pole of the cup where substantial draw-in of the material occured.

Another concept evaluated to explore the deep-drawing capability was reported by Al-Naib and Duncan (23). This method utilizes high pressure oil around a blank perifery to aid in the drawing, and is actually a combined extrusion and drawing process. In this study the Sn-Pb eutectic was used which permitted processing at ambient temperature. Good control of wall thickness was achieved, but the applicability of the process to alloys requiring high temperatures has yet to be demonstrated.

SPF/DB Processes

Recent developments have demonstrated that a series of unique processes are available if joining methods, such as diffusion bonding (DB), can be combined with SPF, and these processes are generally refered to as SPF/DB processes (12,24,25). While DB is not a sheet metal process, when it is combined with SPF a dramatic extension of SPF can result such that a discussion of SPF is incomplete without its inclusion.

The SPF/DB processes have evolved as natural combinations of the SPF and DB processes since the process temperature requirements of both are similar. The low flow properties characteristic of the superplastic alloys aids the DB pressure requirements, and it is found that many superplastic alloys can be diffusion bonded under pressures in the same low range as that used for SPF processing (ie., of the order of 300 to 500 psi). The SPF method used with SPF/DB to date is that of blow forming.

The resulting SPF/DB process consists of the following variations: 1.) Forming of a single sheet onto pre-placed details followed by diffusion bonding shown in Figure 12, 2.) diffusion bonding of two sheets at selected locations followed by forming of one or both into a die as shown if Figure 13 (the reverse sequence can also be used), and 3.) diffusion bonding of 3 or more sheets at selected locations under gas pressure followed by expansion under internal gas pressure which forms the outer two sheets into a die, and, in the process, the center sheet(s) is stretched into a core configuration as shown in Figure 14.

In order to develop diffusion bonds in predetermined local areas, a couple of different techniques has been used. One of these is to use a parting agent, or "stop-off" material, between the sheets in the local areas where no bonding is desired. Suitable stop-off materials may depend on the alloy being bonded and the temperature being used. For example, yttria or boron nitride have been succesfully employed to stop-off titanium alloys processed to temperatures of at least 930°C. Such stop-off materials can be suspended in an appropriate binder such an acrylic. After a DB operation, the area of the stop-off pattern is not bonded, and gas can be applied internally along this pattern thereby causing the external sheets to be separated and formed by expanding into a surrounding die.

A modification of the above method is to use a minimum of four sheets to make a sandwich panel, expanding the external (eg., skin) sheets first, then then bonding (or welding) the inner two sheets to define the core structure, and finally expanding the core to bond with the external sheets and thereby complete formation of the sandwich structure. This sequence is shown in Figure 15.

Forming Equipment and Tooling

The forming of superplastic sheet materials involves methods that are generally different that those used in other more conventional sheet forming processes, and the forming environmental conditions are different. For these reasons the equipment and tooling used are generally different.

Forming Equipment

For the blow forming and vacuum forming methods, there is a need to provide constraint to the forming tools in order to counteract the forming gas pressure. Also, a seal is generally required at the interface between the sheet and the tool around the perifery in order to prevent leakage of the gas pressure. A press is typically used to meet these requirements. Hydraulic presses and mechanical clampling systems have been used, and each has advantages and disadvantages. The hydraulic press can be loaded and unloaded fairly rapidly, but requires a significant capital investment. The mechanical clamping systems are much less expensive, but are more cumbersome to load and unload. Recently, robotic systems have been coupled with a hydraulic press to aid the loading and unloading, and this type of advanced system is especially beneficial for high temperature forming operations such as Ti alloy SPF processing.

The hydraulic presses used include both single-action and multiple-action systems (12,13,20). In the single action press, the press acts to apply the constraining pressure only. In the multiple-action press, the press can also move dies into the forming sheet and effectively aid in the control of the thinning gradients as shown in Figures 9 and 10.

The heating system used must be tailored to the temperature required and the allowable thermal gradients. The most common heat source is that electrical heating, in which resistance heating elements are embedded in ceramic or metal "pressure plates" placed between the tooling and the press platens. This provides for good control of the temperature, and a clean source of energy. The heating platens can be arranged in sections of heating elements, and each section can be controlled by independent temperature controllers to minimized thermal gradients in the forming die assemply. Significant thermal gradients can lead to excessive thinning or rupture of the sheet during forming.

Tooling Materials

The tooling used in the SPF process is generally heated to the forming temperature, and is subjected to internal gas pressure and pressing clamping loads. The internal gas pressure is typically less than about 500 psi and this is usually not the critical design factor for SPF tools. More important are the clamping loads and thermal stresses encountered during heat-up and cool-down and the environmental conditions. The thermal stresses can cause permanent distortions in the die, and this is controlled by selection of a material which has good strength and creep resistance at the forming temperature. Slow heating and cooling of the tooling can reduce the thermal stresses. Materials with a low coefficient of thermal expansion and those that do not undergo a phase transformation during heating and cooling are prefered for the high temperature SPF processes.

The environmental conditions can be severe for the forming of high temperature materials, such as the Ti alloys, Fe alloys, Ni alloys, and other high temperature metals. Oxidation can alter the surface condition of the tooling, thereby affecting the surface quality of the SPF part produced, and eventually affecting dimensional characteristics.

Another important environmental factor is that of compatability between the superplastic sheet and the tooling, and the compatability of these with stop-off materials which may be used. Interdiffusion at this interface between the tooling and the sheet can result in degradation of both of these materials. Reactive metals, such as Ti alloys, are especially prone to this type of problem. Tooling materials that have been found to be successful with Ti alloys are the Fe-22Cr-4Ni-9Mn alloy and similar materials. Parting, or stop-off, agents are also helpful in minimizing the interaction, and materials such as boron nitride and yttrium oxide have been successfully used. Generally, materials with a low solid solubility in the sheet are good canditates for compatability.

A variety of types of materials have been used for SPF tooling, including metals and alloys, ceramics, and graphite (26). The metal tools are preferred for large part production, such as 100 parts or more. The graphite tools are suitable for about 100 parts, and are readily hand worked although there is a problem with shop cleanliness with this material. Ceramics can be cast into the shape desired, and are therefore inexpensive for a variety of large parts. The ceramic is subject to cracking and rapid degradation, and therefore requires frequent repair. Ceramic tools are therefore considered for small production quantities, usually less than about 10 parts.

Thinning Characteristics

In order to take advantage of the very high elongations possible with superplastic metals, it is necessary to accept corresponding significant thinning in the sheet material which is a natural consequence of the deformation conditions. For superplastic deformation, elastic strains are negligible, and therefore constancy of volume can be assumed. From this consideration, the sum of the plastic strains is zero, and tensile strain in one direction must be balanced by compressive (negative) strains in another. The strains are:

$$\epsilon_1 + \epsilon_2 + \epsilon_3 = 0$$

where ϵ is the strain, and the subscripts indicate the principle directions. For example, in a sheet forming operation under plane strain conditions, $\epsilon_2 = 0$ and $\epsilon_3 = -\epsilon_1$. In this case the thinning strain (^g. ϵ_3) is equal and opposite to the longitudinal tensile strain, and the thinning will therefore match the tensile deformation. For large tensile strains, the thinning will be correspondingly large. Accordingly, as the thinning increases, the tendency to develop thinning gradients also increases.

While the superplastic materials are effective in resisting the necking process, they nonetheless do neck (in relation to the m value), and thinning gradients do develop. Therefore, in the design and processing of superplastic formed parts, it is important that the thinning be understood and considered.

Thinning in Uniaxial Tensile Test

It has been shown by a number of investigators (2,5,6-8) that superplastic deformation occurs when m is large, and under these conditions the deformation process is predominately post-uniform, in contrast to conventional metal tensile behavior. In most cases virtually all deformation is non-uniform, and the issue in the tensile behavior is the extent of this nonuniformity. The thinning in the tensile specimen can be assumed to be the result of a pre-existing inhomogeneity (7,8) which can grow under the imposed deformation.

The rate of thinning in the tensile specimen is therefore determined by the size of the inhomogeneity, but also the m value. This has been shown analytically (7) for an idealized tensile specimen as shown in Figure 17 containing a geometric inhomogeneity, f (eg a machining defect). This analysis analytically follows the strain development both inside and outside the inhomogeneity, assuming that the applied load is fully transfered along the length of the specimen, and the material obeys the following constitutive equation:

$$\sigma = K\epsilon^{n}\dot{\epsilon}^{m}$$

where n is the strain hardening exponent (n is small in this case). The results of calculations using this model are shown in Figure 18 where the strain in the inhomogeneity is graphed as a function of the strain outside the inhomogeneity for a number of different m values. The extent of the thinning in the tensile specimen is shown to be strongly related to the m value, although at all m values, thinning gradients will develop if the strain is sufficiently large. It will be seen that this is also the case for the sheet forming where the inhomogeneity is caused by stress gradients resulting from the part geometry and tool interactions. The development of the inhomogeneities in tensile specimens have also been shown to relate to the m value, as shown in Figure 19 for the Zn-22Al eutectoid alloy (27). In this figure, the results are presented for the same alloy tested at different strain rates for which the m value are known to differ.

Thinning in Spherical Domes

While the thinning in superplastic tensile test specimens is the result of geometric inhomogeneities, the corresponding thinning in biaxially formed parts is usually the result of local stress state differences initially which then lead to the development of geometric inhomogeneities. In all of these cases, however, it is the difference in the local stresses which lead to strain rate gradients, and it is the strain rate gradients which directly develop into thickness gradients. A major difference between the tensile specimen and the part configuration is that, in the former, the stress gradients may be varied (ie., reduced) by dimensional control during machining. In the part forming, however, the configuration determines the stress state, and that is not adjustable without changing the geometry.

The concept of thinning during SPF processing is perhaps best understood for the case of the bulging of a sheet (15,16,20,28-34). In this geometry, there is a stress state gradient from the pole of the dome to the edge as shown in Figure 20. If the dome is assumed to develop into part of spherical symmetry, the stress state can be readily described. At the pole, the orthogonal stresses are equal, and the stress state is that of equibiaxial tensile. At the edge of the dome, there is constraint around the perifery leading to a plane strain stress state. Since the flow behavior of superplastic metals has been found to obey the von Mises criterion (35), it is helpful to examine the effective stress, $\bar{\sigma}$, which will determine the corresponding strain rate:

$$\bar{\sigma}_p = \frac{1}{\sqrt{2}}[(\sigma_\theta - \sigma_\phi)^2 + (\sigma_\phi - \sigma_t)^2 + (\sigma_t - \sigma_\theta)^2]^{\frac{1}{2}}$$

If it is assumed that the through-thickness stress is small with respect to the in-plane stresses, the effective stresses at the pole and the edge can be expressed in terms of the meridianal strain, ϵ_θ, as follows:

$$\bar{\sigma}_p = \sigma_\theta$$

and

$$\bar{\sigma}_e = \frac{\sqrt{3}}{2}\sigma_\theta = 0.87\sigma_\theta$$

where σ is the stress, θ, ϕ and t refer to the meridional circumferential, and thickness directions respectively. Therefore, the pole is experiencing a 15% higher flow stress than the edge, resulting in a higher strain rate, the initial magnitude of which depends on the m value.

The stress state difference between the pole and the edge of the dome is roughly the equivalent of the tensile specimen which has a local geometric inhomogeneity, f, of 0.13. The initial strain-rate difference between these two areas is dependent on the m value: the larger the m the smaller the strain rate difference and the less the tendency to develop a thickness gradient. For example, the ratio of $\dot{\epsilon}_e/\dot{\epsilon}_p$ is 0.87 for m = 1 and the value is 0.5 for m = 0.5, both for the same initial effective stress difference.

The stress gradient in a forming dome therefore causes a more rapid thinning rate at the pole, and it may be expected that the thinning difference will accelerate with time, leading to a thickness gradient in the formed dome. There are abundant experimental results to show that this is the case, and that the thinning gradient is a function of the m value. Profiles of thickness for bulge-formed sheets are shown in Figure 21 for m values of 0.57 and 0.23

(28). The thickness gradient is in agreement with expectations, and the effect of the high m value to impede localized thinning at the pole can be seen. Other results are shown for a titanium alloy and a stainless steel for which m values are 0.75 and 0.4 respectively in Figure 22 where the thickness strain is plotted as a function of the position along the dome cross section (16). The position along the dome is measured as the fractional height, h/h_o, where h is the full height of the dome and h_o is the height on the dome at which the thickness measurement is made.

A number of analytical developments have been reported which predict the thinning for superplastic forming this type of geometry (16,20,30,32-34). These models result in relations for thicknesses which are not closed-form, but require numerical integration of strain increments. While they are somewhat cumbersome, the models do predict the thinning characteristics reasonably well, as can be seen by the comparison of experimental and analytical data in Figure 23.

The theoretical predictions can be utilized to clearly show the influence of the strain rate sensitivity of flow stress on the thinning gradient. For example, the thinning for a hemisphere formed from materials of differing m values is illustrated in Figure 22. In this figure, the thinning factor, $s/\bar{s}$, is plotted as a function of the fractional height, where s is the local thickness and $\bar{s}$ is the average dome thickness. The maximum thinning occurs at the pole due to the stress state as mentioned previously, and the strain rate sensitivity is a crucial parameter in determining, not only the initial strain rate difference, but also the subsequent rate of thinning as shown in Figure 24 where the thinning factor at the pole can be seen to be increasingly influenced by m as the dome height is increased.

The initial stress state differences and the corresponding strain rate differences along the meridian of a forming dome lead to a predictable thinning gradient in this type of geometry. The magnitude of the thinning gradient, however, is determined by the strain-rate sensitivity, m, and the height to which the part is formed.

Thinning in Rectangular Shapes

The factors contributing to the thinning characteristics in rectangular-shaped parts, as well as other shapes, are the same as those for the spherical dome-shaped parts discussed above. It is the specific geometry that determines the initial stress state gradients, and different geometries will be expected to develop different stress states in the forming part.

The rectangular shape is one that is common to many parts, or sections of parts, and it has therefore been studied by both experimental and analytical methods similar to those of the spherical dome (12,17,35). For the long rectangular shape, there is a plane stress state throughout the width of the sheet; and for the case where the die entry radius does not cause a significant stress concentration, the sheet will not experience an initial stress state gradient. This case is very similar to that of the tensile test in which only thickness variations, or material inhomogeneities, will cause local stress differences leading to localized thinning. Since these are small in comparison to the magnitude of the stress variation in the forming dome, it may be expected that the thinning gradients would be less pronounced. Experimental results for the free forming cylindrical section show this to be the case, as shown in Figure 25a and 25b, and virtually no thinning gradient is seen for hemi-cylindrical shapes.

Interactions with the tooling do, however, cause local stress variations which can lead to thinning gradients shown in Figure 25c and 25d. This effect can be considered as two different types resulting from differet areas of the die: 1.) the die surface at the bottom and side-wall, and 2.) the die entry radius.

If we ignore for the moment the die entry effects, the die surface can be considered to restrict deformation in the forming sheet where contact has been made and where friction is non-zero. If the friction is large, the forming characteristic is as illustrated in Figure 26. When the sheet make contact with the die wall surface, the deformation in that contact area is restricted, and thinning is localized in the non-contact areas leading to a greater degree of thinning in the last area to contact the die than in first areas to make contact as illustrated. This results in a thickness gradient as shown in Figure 27, for a Ti alloy part formed in a die with no lubricating compounds present.

This type of thinning is readily predicted analytically if it is assume that the sheet "sticks" to the die surface after contact is made by using an incremental method (17). Results of this type of model show that there are a variety of thinning variations corresponding to various width and depth ratios of the rectangular shape as shown in Figure 28. It is apparent from this figure that the narrow and deep parts develop the greatest amount of thinning. It is interesting to note that in this specific case, the thickess profiles can be predicted quite well without referring to the strain rate sensitivity, m. This is the result of the dominant effect of the die friction coupled with the uniform initial stress state.

If the interfacial friction is reduced, the thinning gradient will be reduced in the side-wall and bottom areas, since continued deformation after die contact is possible. An example of the thinning in a formed rectangular Ti part is shown in Figure 29 for which forming was conducted with a boron nitride solid lubricant (12,17).

A die entry radius causes a local stress concentration in the forming sheet which then creates a stress state gradient in the forming sheet, and this can lead to localized thinning, especially if the die radius/sheet thickness ratio is small and if the surface is lubricated. The source of the stress concentration is the back-pressure exerted by the die radius on the forming sheet, and the gas pressure on the opposite side of the sheet from the die. The pressure exerted by the die radius has been shown to be (17):

$$p_r = \frac{\sigma_w h}{R_1}$$

where p_r is the pressure of the die entry radius, σ_w is the in-sheet stress in the width direction, h is the sheet thickness, and R_1 is the die entry radius. This and the applied gas pressure, g, develop an average through-thickness stress σ_h of:

$$\sigma_h = \frac{g + p_r}{2}$$

The magnitude of σ_w will be dependent on the local friction coefficient, μ_r, and the position on the radius, so that the effective stress will vary around the die entry radius. A detailed analytical model of this somewhat complex condition is presented elsewhere (17), but it has been shown that a local stress increase is developed in this area causing a tendency to thin locally; and if the friction is sufficiently low, the initially thinned section can continue to thin after die contact is made. Thus, significant localized thinning can occur, and even rupture may take place if the conditions are sufficiently severe.

Thinning over the die entry radius is the result of stress gradients, and therefore the strain rate sensitivity exponent, m, is an important parameter in determining the extent of thinning that will develop. The influence of these variables is illustrated in Figure 30 for a Ti alloy part formed under the indicated conditions of lubrication and strain rate. The strain rate variations resulted in corresponding variations in the m value during the respective forming process. The thinning for the unlubricated part is in agreement with that expected from the discussion in the above section. For the case where lubricant is used, the strain rate, which determines the corresponding m value, is a factor in determining the extent of thinning over the die entry radius. The average m value corresponding to the die entry radius was higher for the forming process which developed the lower average strain rate, resulting in significantly reduced tendency to locally thin in that area.

Thinning Control

Since superplastic formed parts typically are stretched to very large elongations, the thickness variations are potentially large for a part. It is therefore often an important issue to control the thickness variations in order to meet part tolerance requirements. While it is seldom possible to prevent thickness variations, there are techniques which can be utilized to control this problem. In addition to such methods, the designer can often accommodate variations in thickness if he knows what they may be in advance. This latter approach is an important and viable one, but will not be addressed here since it is considered to be beyond the scope of this paper.

The methods to control thinning are: 1.) processing of the superplastic material to achieve high m value, 2.) use of surface lubrication as discussed above, 3.) use of thermo-forming methods to control the localized deformation, 4.) modification of the die or part design to minimize local stress concentrations, 5. forming a thickness-profiled sheet and 6.) application of pressure in a controlled and profiled manner to control strain rate to a value corresponding to a high m value. Since the raw sheet material is generally obtained from a commercial supplier, the material superplastic properties are under control of the mill. However, it may be judicious for the forming plant to obtain material under control of an appropriate specification. The effect of lubrication was discussed above, along with the effect of the die entry radius which may, in some parts, be increased to minimize the thickness gradients.

The thermo-forming method (14,20) has been shown to offer effective techniques which can control the thinning gradients in single-pocketed deep-drawn parts. With these methods, a moveable tool is usually used to contact the forming sheet before the finished shape is produced, causing the local friction to minimize deformation in some locations while free forming sections continue to deform.

Examples of the use of thermo-forming techniques was demonstrated by Johnson et al (20) using an apparatus as shown in Figures 31 and 32. The test rig used by these investigators consisted of two cylindrical chambers, each 190 mm (7.5 in) inside diameter by 178 mm (7 in) deep, and a hydraulic ram positioned in the bottom chamber which was capable of moving up and down. The material used was Zn-22Al-1.5Cu sheet 1.27 mm (0.50 in) thick. For the convex upward die, the deformation was restricted in the center of the sheet and concentrated at the outer area, resulting in a strain and thickness profile as shown in the Figure 33 where the top center can be seen to be thicker than the adjacent areas. This thickness profile was substantially modified by the use of a concave upward die as shown in the Figure 34. In this case the superplastic diaphragm is formed down into the concave die by gas pressure, and the die was slowly withdrawn until it reached the bottom. The pre-formed diaphragm was then formed into the upper cylindrical chamber in the same manner as that of the previous figure. The resulting profile is seen to be considerably more uniform across the top of the part.

The use of thickness profiled sheet is one that has been suggested to control the thickness in the final part (32). The concept considers that the initial thickness variations can be used to offset the subsequent variations resulting from the stress state and part geometry effects on the thinning as shown in Figure 35. If areas which will thin excessively are thicker than surrounding areas, it is possible to develop finished part thickness profiles that are more uniform than those formed of constant thickness sheet.

Pressure Profiling

It is now well recognized that the m value for superplastic alloys will vary with strain rate, and often it will vary also with strain. The strain rate imposed during the forming process will therefore determine the m value, and if the strain rate varies during the forming process, the corresponding m value and related thinning uniformity will also vary. For example, the simplest pressurization concept for SPF processing is that of constant pressure. The resulting strain rate for a spherical dome part configuration has been shown (15) to be as much as three orders of magnitude. A graph of the predicted strain rate for a sherical dome which will be uniform generated under constant pressure for such a part is shown in Figure 36. The strain rate drops to 0.001 of the initial value when the part becomes a hemisphere. As can be seen in Figure 3, the m value for a typical superplastic alloy can vary from a maximum in m to values less that 0.2 over strain-rate ranges of this magnitude. The consequence of forming a part under these conditions is that excessive thinning or part rupture during forming are likely. While the strain rate corresponds to a large m, good thinning resistance is maintained. However, this would be transient, and other strain rates would be encountered corresponding to low m values, and poor resistance to localized thinning would be present.

This condition can be rectified if the strain rate can be maintained at a constant level corresponding to a suitably high m value. Since the constant pressure is seen to develop a variable strain rate, it is apparent that a variable forming pressure would be required to develop a constant strain rate. Such pressure profiles have been established analytically for the spherical dome (15,16) and the rectangular shaped (17) part configurations. Since most of the analytical models of the SPF process utilize the applied gas pressure to establish the current stress and strain-rate conditions, it is possible to utilize these same models to adjust the current gas pressure to develop the desired stress and strain rate.

The resulting pressure profiles for constant strain-rate forming of the spherical dome and the rectangular parts are *illustrated in Figures 37 and 38 respectively. It is typical* that the pressure initially rises rapidly followed by decrease. The rapid initial rise is due to rapid decrease in the radius of curvature with little change in thickness, and the subsequent decrease is the result of thinning which is more rapid at this stage than the change in radius of curvature. The depth to which a part is formed will also affect the pressure profile for constant strain rate control, as shown in Figure 39 for a rectuangular part with formed with no lubrication. For a shallow part (W = 12in), the applied pressure never reaches a maximum, but continues to rise during the forming process. For a deep part (W = 2in), the pressure is decreased for a significant time period before being increased to high levels.

Cavitation and Cavitation Control

Most superplastic alloys tend to form voids, or cavities, at intergranular locations during the superplastic deformation. This process is termed "cavitation". The cavitation can lead to degradation of strength and other design properties, and is dealt with in one of two ways: 1.) establish reduced design properties or 2.) utilize a "back-pressure" technique to control cavitation.

Typical cavitation as a function of strain is shown for an aluminum alloy in Figure 40. It can be seen that the absolute amount of cavitation in terms of the volume fraction is not large but depends on the strain imposed. The use of the back pressure concept imposes a hydrostatic pressure on the sheet during forming, and if this pressure is of the order of of the flow stress, cavitation can be reduced or completed suppressed. An example of the effect of back pressure on the rate of development of cavitation is shown in Figure 41.

In practice, the back pressure is achieved by imposing a pressure on the back side of the sheet, to oppose the forming pressure, and sustain this pressure during the forming cycle. The forming pressure must be higher that the back pressure, and the same forming rates can be achieved with as without back pressure if the pressure differential is the same. For example, if a pressure profile is desired, such as that shown in Figure 37 or 38, the pressure profile is simply raised in magnitude by an amount equal to the back pressure. Since the back pressure is normally of the order of the material flow properties, pressures of about 100 to 500 psi are generally suitable to suppress cavitation.

Summary and Conclusions

The SPF process is unique in terms of the complexity of parts that can be produced and the methods which can be used to shape such a material. A number of processing methods are currently being used, most of which involve significant stretching of sheet material. The high ductility which can be achieved with these types of materials also has a consequence that must be understood and dealt with, ie., that of thinning gradients. The thinning gradients are a natural consequence of the stress gradients that develop in the various die configurations, and it is the superplastic property of strain rate sensitivity of the flow stress that then determines the subsequent thinning gradient that will result in the part. Control of the forming process, die configuration, and material characteristics are all factors which can affect the thinning.

The SPF processes are being using increasingly for a wide range of structural and non-structural applications. The availability of superplastic alloys is considered to be a major impediment to the broader use of these processes, but it may expected as the use of the technology increases, more and perhaps better superplastic alloys will become available.

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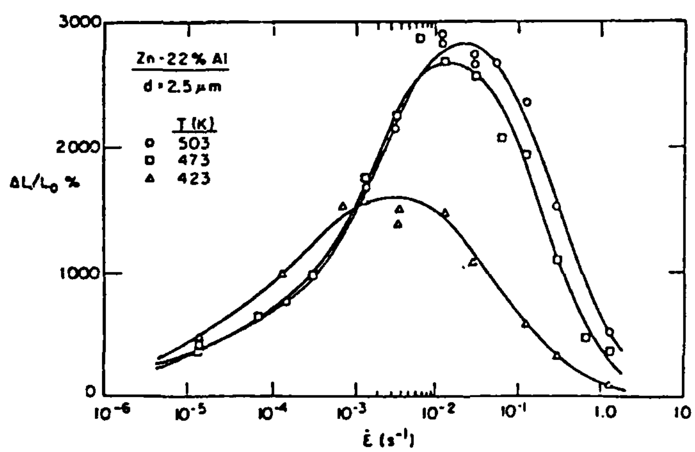

Fig. 1. Tensile fracture strain vs. initial strain rate for a Zn-22Al alloy having a grain size of 2.5μm tested at temperatures in the range of 423 to 503 K.

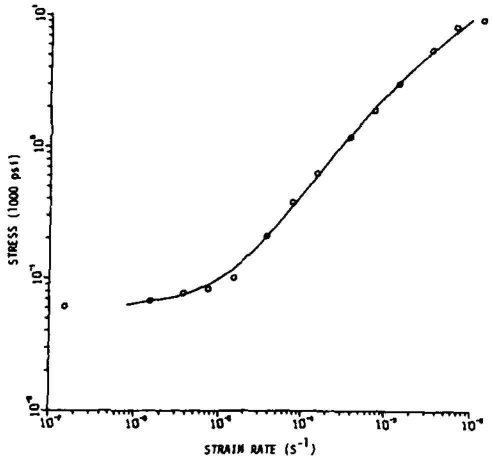

Fig. 2. Log stress vs. log strain rate for Ti-6Al-4V at 927 C.

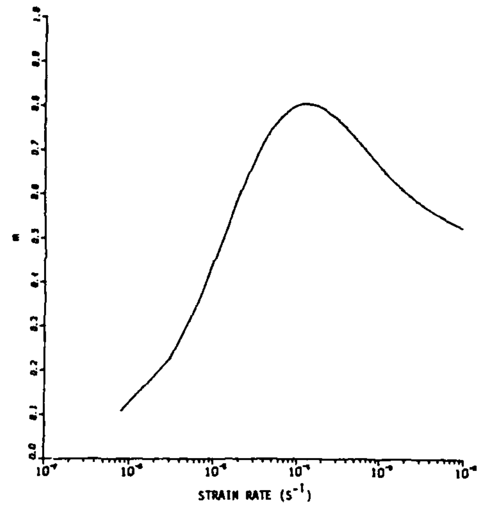

Fig. 3. m vs. log strain rate for Ti-6Al-4V at 927 C. This curve corresponds to the data shown in Fig. 2.

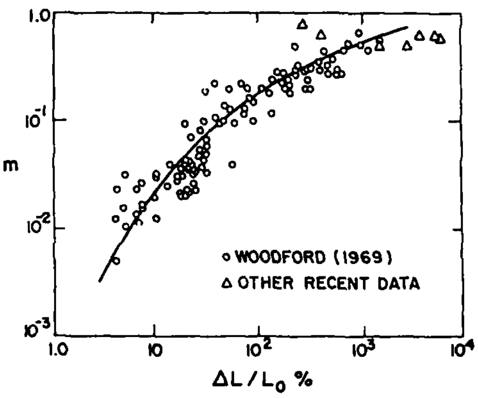

Fig. 4. m vs. elongation for a number of alloys (5).

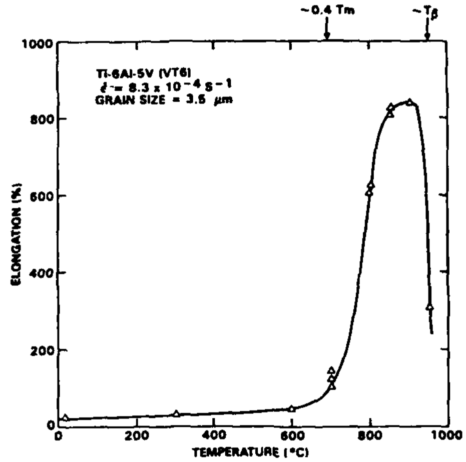

Fig. 5. Elongation as a function of temperature over a temperature range including superplastic deformation, for the Ti-6Al-4V alloy tested at a strain rate of 8.3×10^{-4} (9).

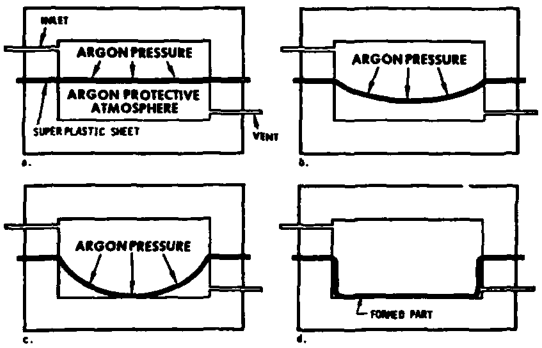

Fig. 6. Illustration of the blow forming technique for superplastic forming. The sequence a. through d. indicates the progression of forming with increasing time.

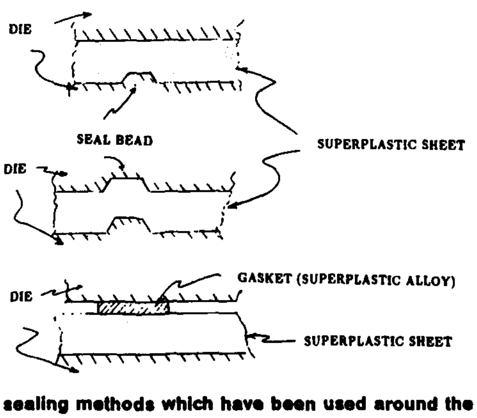

Fig. 7. Illustration of various sealing methods which have been used around the sheet to provide for a pressure seal suitable for containing the gas pressure during forming. The sections a. and b. utilize "seal beads" machined into the tooling, and c. shows the use of a superplastic frame used as a soft gasket.

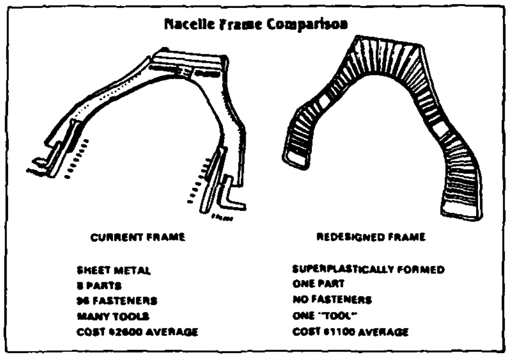

Fig. 8. An aircraft nacelle frame which was redesigned from a conventional configuration to one suitable for superplastic forming and one which uses fewer parts and fasteners. The part was fabricated from Ti-6Al-4V.

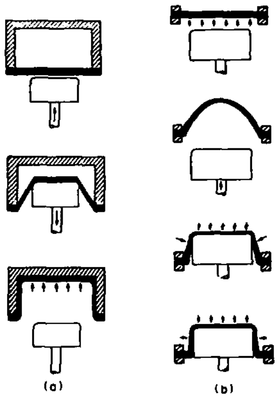

Fig. 9. Examples of thermo-forming methods used for superplastic forming, involving a. "plug-assisted forming" into a female die cavity, and b. "snap-back forming" over a male die which is moved up into the sheet.

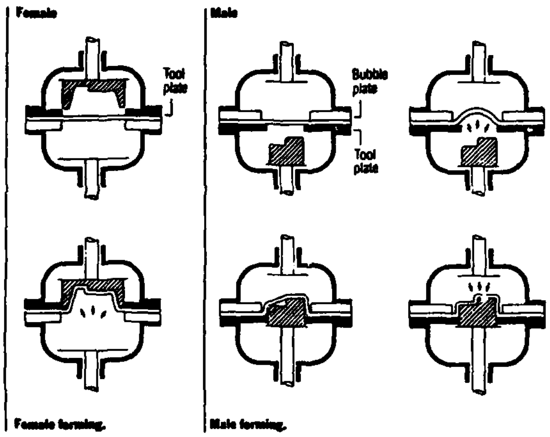

Fig. 10. Thermo-forming methods which use gas pressure and movable tools to produce parts from superplastic alloys. Both female and male movable dies have been employed as shown.

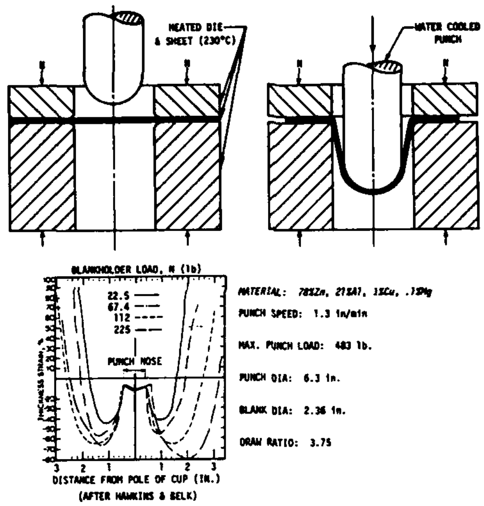

Fig. 11. Sketch of the punch forming set-up for deep drawing a superplastic sheet. Also shown are the thinning profiles resulting from deep drawing with blank holder loads.

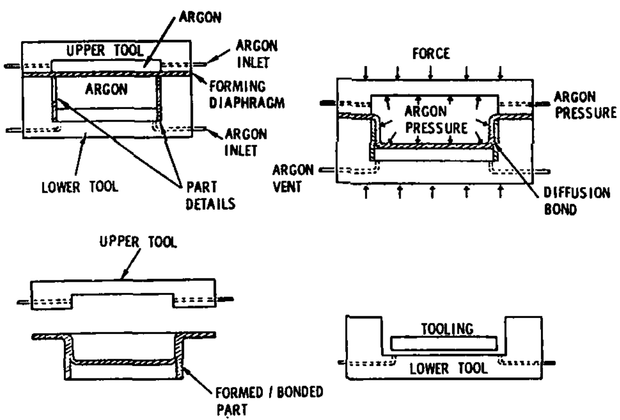

Fig. 12. Drawing of the cross section of the SPF process combined with diffusion bonding (SPF/DB). The process shown utilizes pre-placed details to which the superplastic is bonded.

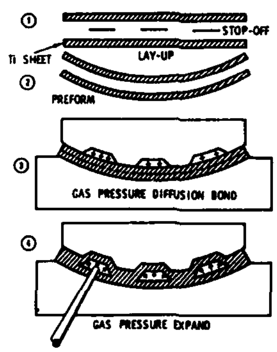

Fig. 13. Illustration of the SPF/DB process for two-sheet parts.

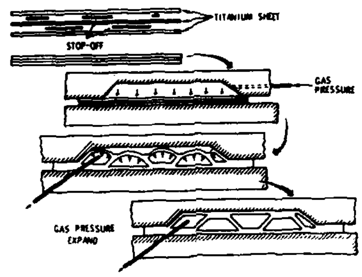

Fig. 14. Illustration of the SPF/DB process for three-sheet parts.

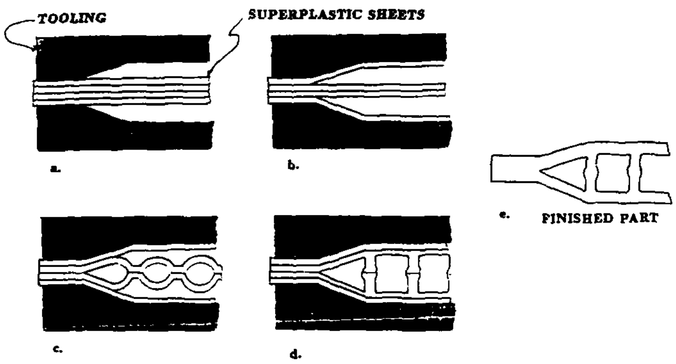

Fig. 15. Example of a four-sheet SPF/DB process in which the outer sheets are formed first and the center sheets are then formed and bonded to the outer two sheets.

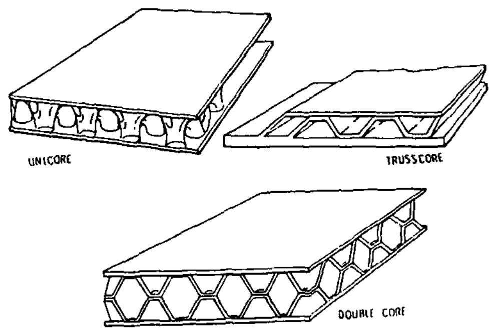

Fig. 16. Examples of sandwich configurations producible by the SPF/DB process using three or more sheets of a superplastic alloy.

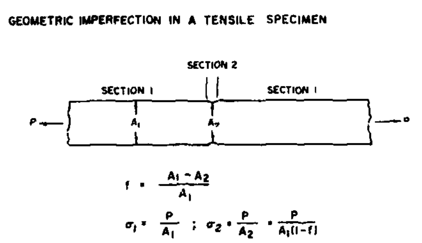

Fig. 17. Geometric inhomogeneity in a tensile specimen.

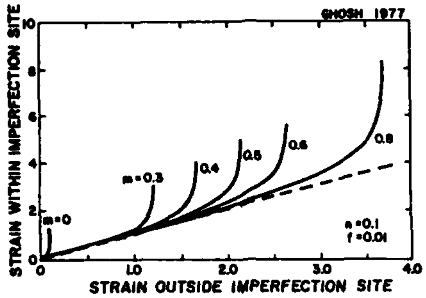

Fig. 18. Calculated strains inside and outside an inhomogeneity in a tensile specimen, such as that in Fig. 17, for various m values.

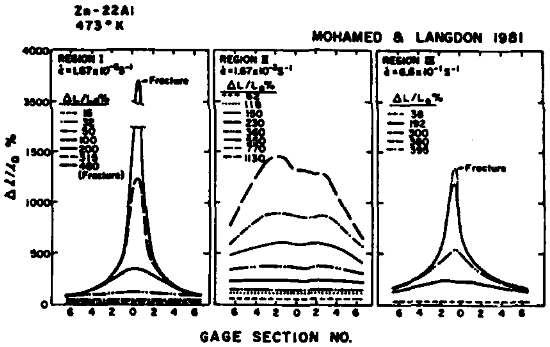

Fig. 19. Local elongation gradients in tensile specimens of the Zn-Al alloy after testing within the superplastic strain rate range (Region II) and outside the superplastic strain rate range (Regions I and III).

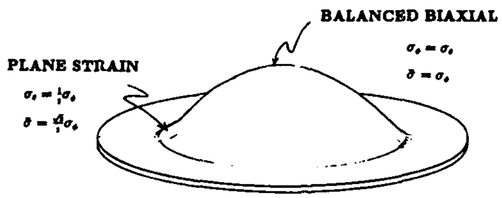

Fig. 20. Sketch of a spherical dome indicating the range of stress states existing between the pole and the edge.

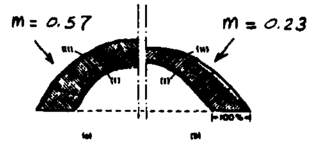

Fig. 21. Experimentally observed thickness profiles for a hemispherical dome formed from materials with different m values.

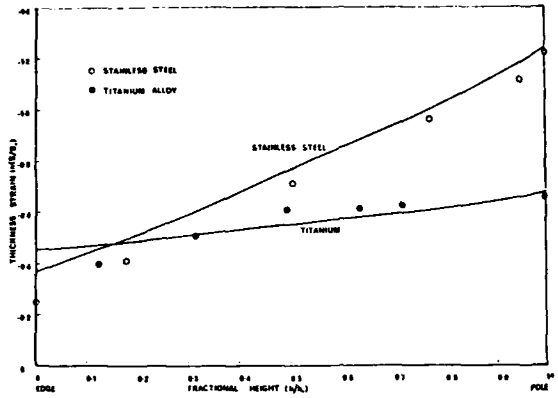

Fig. 22. Thickness strain as a function of the fractional height for dome-shaped parts formed from a stainless steel with an m value of 0.4 and a titanium alloy with an m value of 0.75. (16)

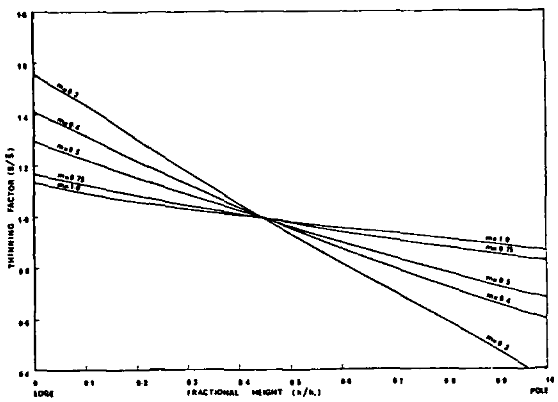

Fig. 23. Theoretical relations for a hemisphere showing the thinning factor as a function of the fractional height for a range of m values for 0.3 to 1.0. (16)

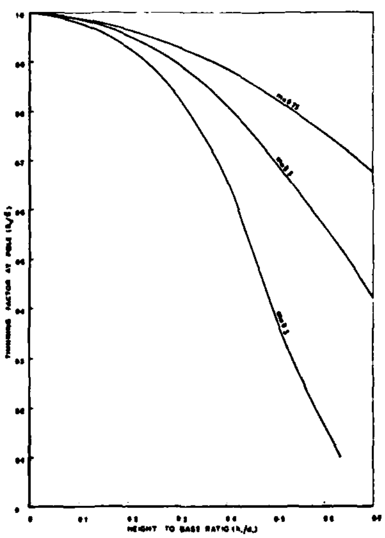

Fig. 24. Theoretical curves showing the thinning factor at the pole as a function of the bulge height-to-base ratio for m = 0.3, 0.5, and 0.75. (16)

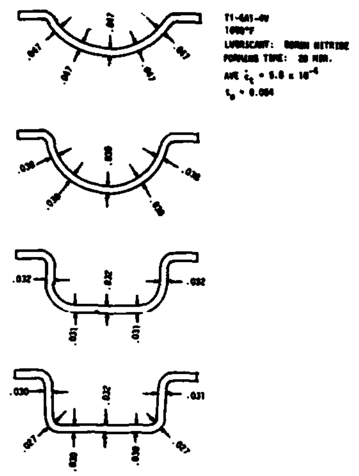

Fig. 25. Thinning development in a superplastic formed Ti-6Al-4V part of rectangular cross section and semi-infinite length.

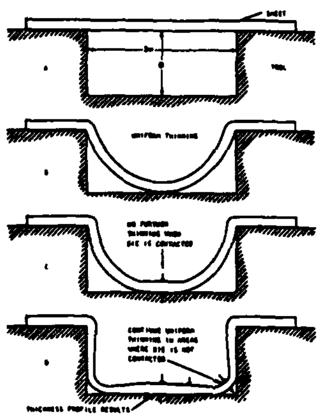

Fig. 26. Illustration of thinning characteristics in the blow forming of an unlubricated part of rectangular cross-section and semi-infinite length.

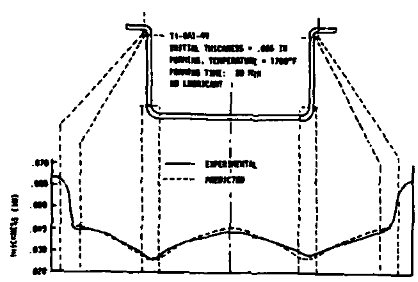

Fig. 27. Observed and predicted thinning profiles in an unlubricated blow-formed Ti-6Al-4V alloy part of rectangular cross section.

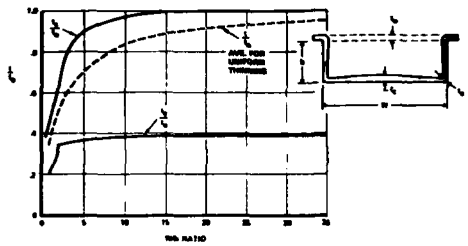

Fig. 28. Predicted minimum thicknesses as function of the width-to-depth (w/h) ratio for unlubricated blow-formed parts of rectangular cross section.

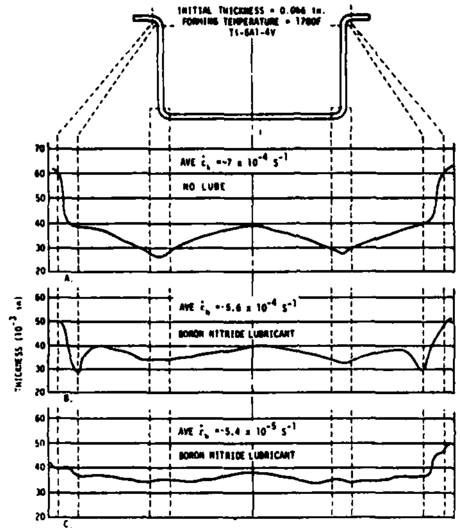

Fig. 29. Observed thickness profiles for blow-formed parts of rectangular cross section formed under different average strain rates and with different lubrication conditions.

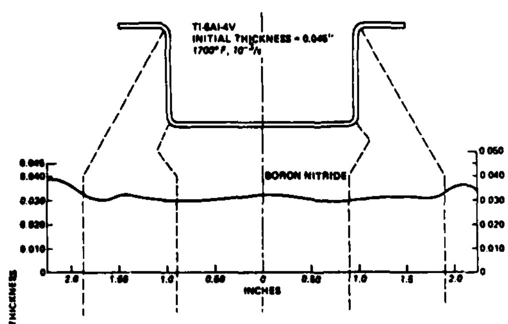

Fig. 30. Observed thickness distribution for parts formed at 10^{-3} s^{-1} with boron nitride lubrication.

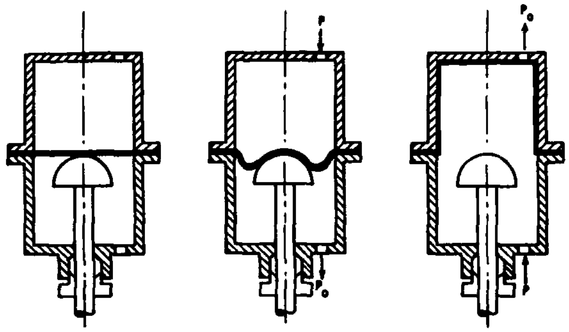

Fig. 31. Apparatus for thermo-forming superplastic sheet materials using a convex die member to control thinning in forming of a "hat" configuration. (20)

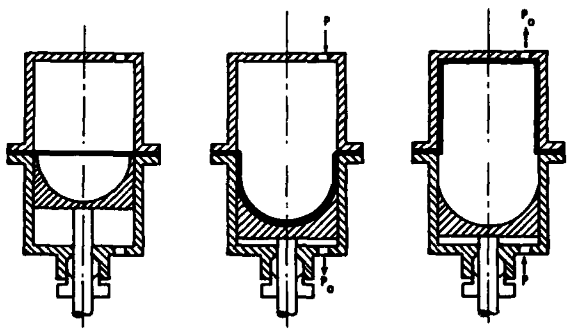

Fig. 32. Apparatus for thermo-forming superplastic sheet materials using a concave die member to control thinning in forming of a "hat" configuration. (20)

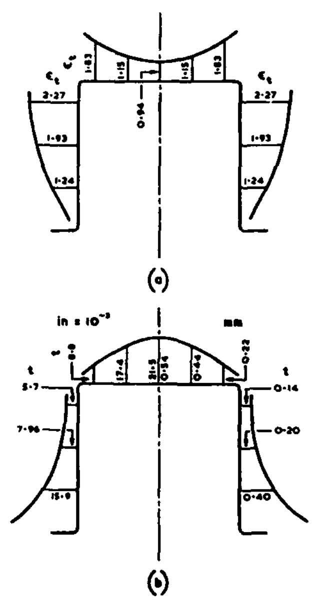

Fig. 33. Thickness profile for "hat" configuration formed with a convex die member as shown in Fig. 31. (20)

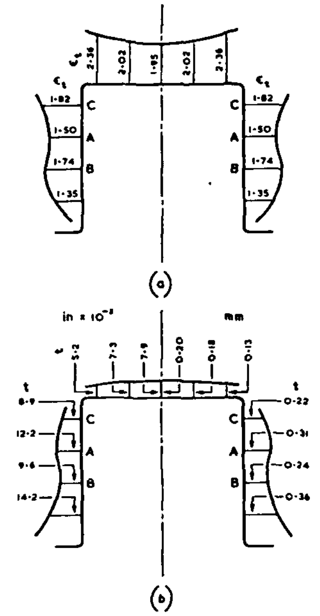

Fig. 34. Thickness profile for "hat" configuration formed with a concave die member as shown in Fig. 32. (20)

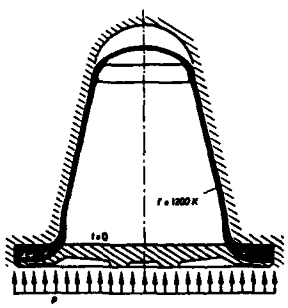

Fig. 35. Sketch of the concept of forming a sheet material which has a thickness profile before forming. (32)

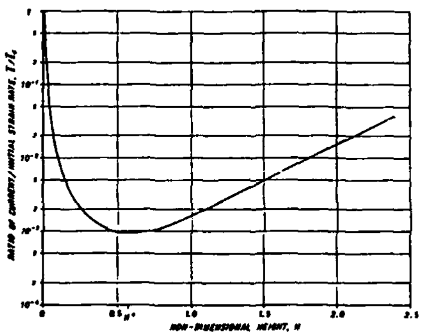

Fig. 36. Ratio of the current to initial strain rate as a function of non-demensional height for a constant pressure application in the forming of a hemispherical configuration. (15)

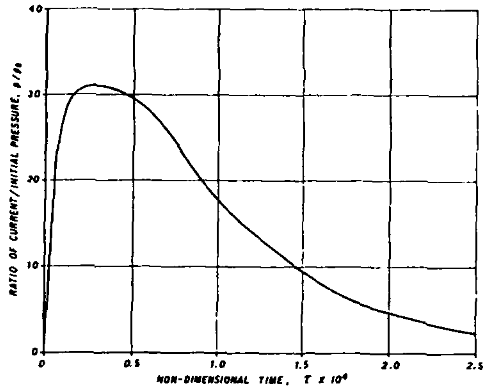

Fig. 37. Ratio of current to initial pressure as a function of a time parameter for forming a spherical configuration under constant strain rate conditions. (15)

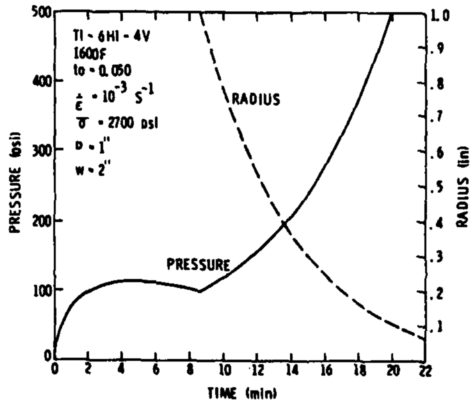

Fig. 38. Analytically predicted pressure vs. time for a constant strain rate in the forming of a long rectangular part. (12)

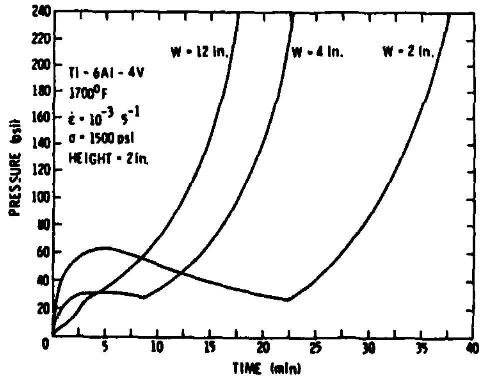

Fig. 39. Analytically predicted pressure profiles for constant strain rate forming of long rectangular parts of different cross section width and height dimensions. (12)

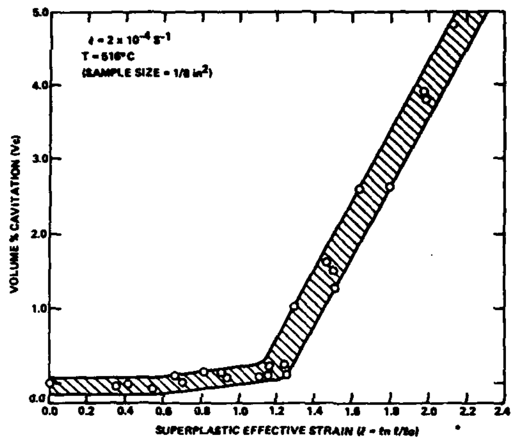

Fig. 40. Development of cavitation with uniaxial tensile strain in a 7475 Al alloy deformed at 516 C under a constant strain rate of 2x10^{-4} s^{-1}.

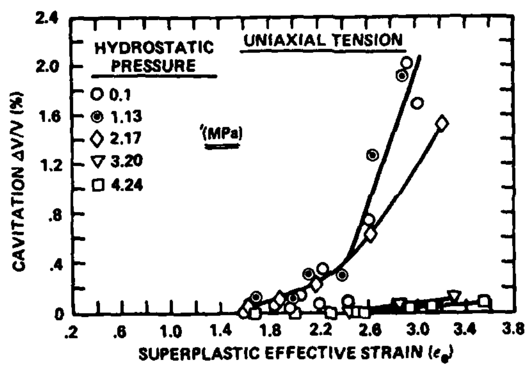

Fig. 41. Effect of hydrostatic pressure on the suppresion of cavitation in 7475 Al superplastically deformed at 516 C.

ADVANCES AND FUTURE DIRECTIONS IN SUPERPLASTIC MATERIALS

A paper prepared for the 1989 NATO-AGARD Lecture Series on Superplasticity

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ABSTRACT

A review is presented of new events in the development and modelling of superplastic materials. Over the last several years, superplasticity has been developed in fine microstructures in microduplex stainless steels, aluminum-lithium alloys, mechanically alloyed aluminum, silicon carbide whisker-reinforced aluminum alloys, aluminum-magnesium alloys, and nickel alloys. In the latter three cases, the phenomenon is observed at high strain rates by comparison with most superplastic alloys. In a major breakthrough, a ceramic (yttria stabilized zirconia) and a ceramic composite (yttria stabilized zirconia containing alumina) have been shown to be superplastic in tension tests. Superplastic behavior in iron carbide has also been observed. Superplastic studies are also underway in intermetallics such as nickel silicides, nickel aluminides, and titanium aluminides. New insights are becoming apparent in the area of modelling superplastic behavior by the integration, into existing models, of new concepts based on grain boundary sliding accommodated by slip and the influence of threshold stresses associated with grain boundary sliding. Newtonian-viscous flow can be approached in fine-grained Class I solid solution alloys. Significant advances have also been made in understanding internal stress superplasticity; recent examples include zinc, alpha-uranium, zinc-alumina composites and silicon carbide whisker reinforced aluminum alloys. A new model has been proposed for internal stress superplasticity. In other major developments, the mechanisms of superplasticity have been successfully applied to consolidation of powders and superplastic behavior has been developed in ferrous-based laminated composites. Finally, some neglected areas offering the potential for superplastic flow are reviewed.

INTRODUCTION

It is often thought that superplasticity is a recently-discovered phenomenon. The history of superplasticity, however, may date back to the early Bronze period (about 2200 BC). For example, Geckinli (1) has speculated that the ancient arsenic brasses, containing up to 10 wt% As, used in Turkey in the early Bronze period, could have been superplastic. This is because these materials are two-phase alloys that may have developed the required stable fine-grained superplastic structure during hand forging of intricate shapes. Furthermore, the ancient steels of Damascus, in use from 300 BC to the late 19th century (2,3), are very similar in composition to modern ultrahigh carbon steels that have recently been developed, in large part, for their superplastic characteristics (4,5).

We believe that 1989 is the seventy-seventh anniversary of the first published report on superplasticity. During the course of our own research into the history of Damascus steels, we came across a paper published in 1912, by Bengough (6), that contains the first description of superplasticity in a metallic material. Bengough describes how "a certain special brass which he had examined pulled out to a fine point, just like glass would do, having an enormous elongation." The quote was made by Bengough in a written discussion following a paper by Rosenhain and Ewen (7) on the "amorphous cement" theory. Examination of the original work by Bengough (6) shows that the special brass was an $\alpha + \beta$ brass and exhibited a maximum elongation of 165% at 700°C. It is of interest to note that rather similar materials have been processed to be superplastic in recent times. For example, Cu-40%Zn brasses have been developed for superplasticity in the temperature range 600-800°C (8). Following the observation of Bengough, only sporadic studies were carried out, as reviewed by several authors (8-10), until the work of Backofen and his colleagues (11) in 1964. At this time, the superplastic characteristics of a Zn-Al alloy were demonstrated and described. The current strong interest in superplasticity can be traced back to this work.

Superplasticity, then, is the capability of certain polycrystalline materials to undergo extensive tensile plastic deformation, often without the formation of a neck, prior to failure. The subject of superplasticity in metal alloys has been reviewed extensively (8-10,12). In certain metal alloy systems, tensile elongations of thousands of percent have been documented. In fact, the subject is of sufficient general scientific interest that a world record has been recognized for the phenomenon in metal alloys (13). Recently, the world record of 4,850% in a Pb-62wt%Sn alloy (14) was exceeded in a commercial aluminum bronze (Cu-10wt%Al based alloy) by a value reported as 5,500% (15,16). Both these samples are shown in Fig. 1. It has recently been brought to our attention that the record has since changed hands twice, with a value of elongation-to-failure of 7,550% now claimed in the Pb-Sn system (17) and, most recently, a value of about 8000% in commercial aluminum bronze (18).

Despite these large elongation values, of many thousand percent, most superplastic alloys exhibit optimum tensile elongations of about 300 to 1000%. This range of values is more than sufficient to make extremely complex shapes using superplastic forming technology (12). It should be recognized that high values of the strain rate sensitivity exponent, m, are also required in order for uniform thinning to accompany high tensile elongations. Large cost and weight savings (through redesign) have provided the driving force in commercial manufacturing for the change from conventional to superplastic forming. The principal alloy systems which have been exploited commercially for superplastic forming are those based on nickel, titanium, and aluminum (12). Interest in superplasticity and in superplastic forming, as measured by the increase in published papers on the subject, is on the increase as shown in Fig. 2. Also shown on the figure are some key events in the recent history of the development of superplasticity.

Since the 1982 International Conference on Superplastic Materials in San Diego (12), a number of other international symposia have been held. Superplastic forming was the topic of a symposium held in Los Angeles in 1984 (19). A conference on superplastic aerospace aluminum alloys was held in Cranfield, United Kingdom, in July 1985 (20), and a general conference on superplasticity was held in Grenoble, France, in September 1985 (21). AGARD has selected superplasticity as one of their lecture series in 1987 (22). Bilateral symposia on superplasticity between China and Japan have been held in Beijing in 1985 (23) and in Yokohama in 1986 (24). In the first Cino-Japan symposium, 32 separate papers were presented over a 4-day period with considerable emphasis placed on superplastic ferrous and aluminum base alloys. Both the Chinese and Japanese Governments have selected superplasticity, in 1980, for intense national research and development studies. They envision superplasticity "as a future technology into the next century." An international conference on superplasticity held in Blaine, Washington, in August, 1988 (25), had a large number of delegates from Iron Curtain countries including five from the Soviet Union. The Soviet delegation was headed by Professors O. Kabaiyshev and O. Smirnov. Kabaiyshev, author of a book on commercial superplastic alloys developed in the Soviet Union (26), is director of an Institute on Superplasticity involving over 200 personnel. Smirnov is program director of a 30-person effort on superplasticity at the Moscow Institute for steels and alloys. Bochvar, son of A.A. Bochvar, who developed the first commercial superplastic alloys (27) based on the Zn-Al monotectoid composition, also attended the conference.

TYPES OF SUPERPLASTICITY

Superplastic materials exhibit high values of strain rate sensitivity, m, in the equation $\sigma = K\dot{\varepsilon}^m$ where σ is the true flow stress, $\dot{\varepsilon}$ is the true strain rate, and K is a constant. Ideal or Newtonian viscous behavior is found in materials where m=1. Most normal metals and alloys exhibit $m \leq 0.2$ whereas superplastic alloys have values of $m \geq 0.4$. There are two well-established types of superplastic behavior in polycrystalline solids. The first type of superplastic behavior is known as fine-structure superplasticity (FSS) and is described in Section A. The second type is known as internal stress superplasticity (ISS) and is described in Section B. In the case of FSS materials, a strain rate sensitivity exponent equal to about 0.5 is usually found and the materials deform principally by a grain boundary sliding mechanism. (The possibility does exist to achieve values of m=1 in FSS through the development of fine grained materials that incorporate glide-controlled slip in the deformation mechanism for accommodation of the grain boundary sliding process.) In the case of ISS materials, however, the strain rate sensitivity exponent is usually unity, i.e., they exhibit Newtonian viscous flow. These ISS materials need not be fine grained, and generally deform by a slip deformation mechanism.

The concepts and principles described in FSS and ISS superplasticity have been applied to enhanced powder consolidation through superplastic flow and the development of superplasticity in laminated composites containing at least one superplastic component. These developments are described in Section C and D.

There are other observations of large tensile strains in metals that do not fit into the above classifications and these are described in Section E. Included are: (a) the observations of superplastic-like behavior (up to several hundred percent) in coarse-grained Class I solid solutions, (b) the possibility of achieving high elongations in relatively coarse-grained materials exhibiting values of m=1 at low strain rates through Coble creep (grain boundary diffusion controlled), Nabarro-Herring creep (lattice diffusion controlled), and Harper-Dorn creep (slip controlled), and (c) the observations of large plastic strains in Cu and Al under the extremely high strain rates found in anti-armor shape charge liners.

SECTION A. FINE STRUCTURE SUPERPLASTICITY (FSS)

Most crystalline materials that are superplastic have this unique property because they are fine-grained. The most common superplastically-formed products are made from fine grained sheets. The principal method is by blow forming, in which gas pressure is applied on one side of a sheet, whereby the sheet plastically flows into a die of predetermined shape and complexity. Recent examples of superplastically formed components include a parachute head box made from an Al-Li alloy of composition 2.6% Cu, 2.4% Li and 0.18% Zr; this alloy is of special interest because of its pre-commercial status as Al-2090 (28). Another area of superplastic forming of sheet materials is that of stainless steel components for sanitary ware in commercial aircraft. The steels are 25%Cr-5%Ni microduplex alloys and these articles are manufactured by Incoform of Birmingham, United Kingdom. A variation of the process of gas pressure forming is the combined superplastic-forming and diffusion bonding process (SPF/DB). Many sandwich structure products are made by this process with fine-grained titanium alloys, usually the Ti-6Al-4V alloy (21,29,30). Diffusion bonding is readily achieved because of the fine-grain size and because the oxide diffusion barrier in titanium dissolves at the temperature of processing, i.e., about 925°C.

It is quite well accepted that superplastic gas pressure forming of fine-grained sheet materials has come of age, and is in "a state of maturity" (31). Another method of superplastically forming parts is by press-forging of bulk material into complex dies. Such processing of ultra-fine grained materials is still in the early stages of development. The best known example of a product made in this manner is a disk with turbine blades made from a fine-grained nickel base alloy (32); an example is shown in Fig. 3. Another possible application of press forging superplastic materials is the production of gears. A bevel gear that has been warm forged from a fine-grained ultrahigh carbon steel (UHCS) is illustrated in Fig. 4 (4). An additional advantage of using superplastic ultrahigh carbon steels is that the carburizing step in normal gear production is eliminated. In UHC steels, the fine structure developed for superplasticity can also lead to considerably better room-temperature properties than are possible with traditional gear steels. Manufacture of die components by superplastic hobbing is another application which is particularly amenable to fine grained high carbon steels. Pearce and Miller (33) have performed investigations in this promising field. Conventional die steels have been made superplastic by thermal cycling heat treatments by Yang Chun-Xiao and his colleagues (34). These steels exhibit high strain-rate sensitivity at 800°C and superplastically-formed dies have been made for use in die casting and in cold extrusion operations.

Very limited quantitative studies have been made in evaluating the variables involved in making bulk products from superplastic materials. It is clear, however, that this field is worthy of much quantitative study. Of special potential significance is the possibility of achieving new bulk shapes through powder metallurgy with the use of superplastic fine-structure powders. Another area related to bulk forming is the preparation of laminated composites based on dissimilar materials wherein one of the materials is superplastic. Examples of powder metallurgy processing under superplastic conditions, and the preparation of laminated composites containing superplastic materials, will be described later in the paper.

Structural Prerequisites for FSS Materials.

The structural prerequisites for developing superplastic materials have been well established for metal-based, fine-grained materials. They are, however, less clearly-developed for non-metallic, fine-grained materials. In the following, a number of prerequisites is given for the development of fine-structure superplastic materials.

Fine Grain Size.

One of the major requirements for fine structure superplasticity is that the grain size, $\bar{L}$, should be small. Typically, the grain size should be on the order of 1 to 5μm. This is because the strain rate increases with a decrease in grain size when grain boundary sliding is the rate-controlling process. Thus, grain size refinement represents a powerful method of increasing the strain rate for superplastic forming of alloys. Another important attribute of achieving a fine grain size is that, for a given rate of deformation, the flow stress decreases as the grain size is decreased. Hence, only low applied forces need to be applied during bulk superplastic forming thereby reducing costs in energy and in die wear.

Presence of Second Phase.

Almost invariably, it is very difficult to observe superplasticity in single phase materials because grain growth is too rapid at temperatures where grain boundary sliding occurs. Therefore, in order to maintain a fine grain size in the superplastic forming range, the presence of a second phase at grain boundaries is required. For this reason, many superplastic materials are based on eutectoid (e.g., $Fe-Fe_3C$), eutectic (e.g., Al-Ca) or monotectoid (e.g., Zn-Al) compositions. These materials can be thermomechanically processed to develop fine, equiaxed, two phase structures. Inhibition of grain growth is usually improved if the quantity of second phase is increased, provided that the size of second phase is fine and its distribution is uniform (e.g., hypereutectoid $Fe-Fe_3C$ alloys). The only ceramic-base materials made superplastic to date are based on the above principle. An example of such a ceramic is yttria-stabilized tetragonal zirconia polycrystal (Y-TZP) consisting of 90% tetragonal phase zirconia and 10% cubic-phase zirconia. Wakai et al. (35,36) showed that this material, when $\bar{L}$ =0.3μm, was superplastic at 1450°C with up to 200% elongation and a strain-rate-sensitivity exponent of 0.5. Apparently, this technology is already being used to manufacture ceramic components. Most recently, studies by Nieh, McNally, and Wadsworth (37,38) on this material have shown elongations of up to 800% in Y-TZP and elongations of up to 500% on ceramic composites based on fine-grained zirconia, e.g., 20% Al_2O_3/Y-TZP. Examples of these results are shown in Fig. 5. Kim et al. (39) have acheived superplasticity in a fine grained iron carbide. Strain rate sensitivity exponents of 0.5 to 0.7 were achieved in the temperature range from 700 to 1000°C at strain rates from 10^{-4} to 10^{-2} s^{-1}. The material was prepared from rapidly solidified powders of a 5.25%C-1.5%Cr hypereutectic iron. After compaction and extrusion at 1000°C, the microstructure consists of 80% of fine equiaxed grains of Fe_3C and 20% of iron. The microstructure is illustrated in Fig. 6 and also shown is a sample deformed to 610% elongation, at 1025°C, and at a true strain rate of 1 x 10^{-4} s^{-1}. The grain size after deformation remains equiaxed and fine ($\bar{L}$ = 4μm). Also, in recent times the subject of superplasticity in intermetallics has started to receive attention. For example, Nieh and Oliver have achieved an elongation of 650% at 1080°C and at a strain rate of 10^{-3} s^{-1} in the Ni_3Si intermetallic (40). Other examples of superplasticity in intermetallics are also emerging. In a recent review by Mukherjee et al. (41), superplastic behavior was reported in Ni_3Al based alloys (640%), Ti_3Al (500%), and TiAl (unspecified elongation). Finally, in two recent reviews on superplastic behavior (41,42) the possible superplastic behavior of geological materials has been described. There is evidence for both fine structure and internal stress superplasticity in these materials. In fine structured geological materials, indirect evidence, such as determination of

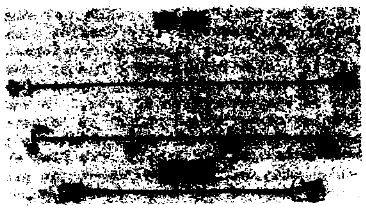

Fig. 1 A superplastic elongation of 4,850% in a Pb-62 wt% Sn alloy (14) is demonstrated by the two top samples and 5,500% is demonstrated by the two bottom samples in a commercial aluminum bronze (16). The current world record is about 8000% in the commercial aluminum bronze (18).

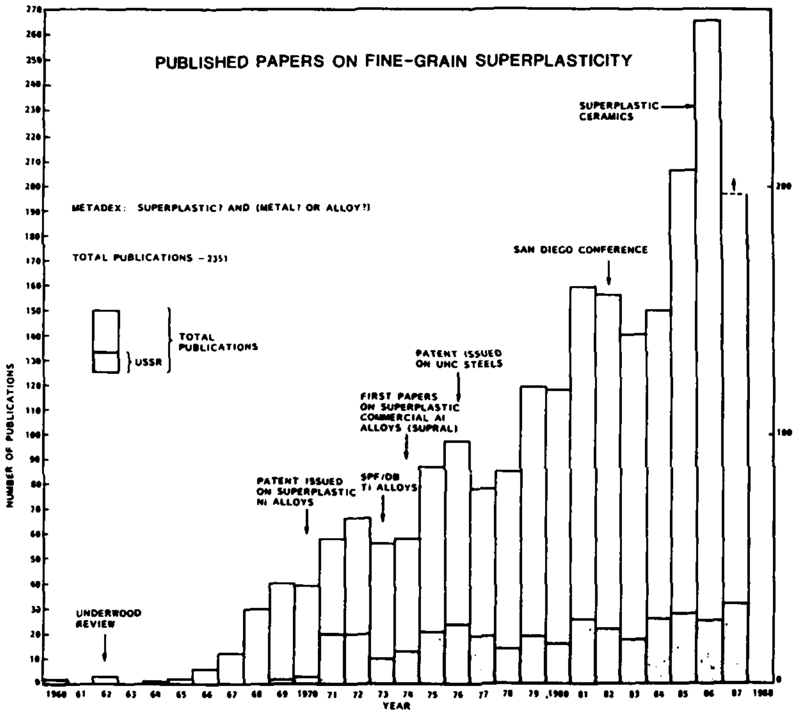

Fig. 2 The number of publications in superplasticity from 1960 to the present time.

Fig. 3 Net shape forming of an ultrafine-grain-size, nickel-base alloy by superplastic forming in two stages. (Left) Original powder metallurgy IN 100 billet. (Center) Powder metallurgy billet pressed into disk shape. (Right) disk-shaped billet superplastically pressed into disk and turbine blades. (Courtesy of J. Moore and R. Athey (32), Pratt and Whitney, Florida [AF Contract F33615-72-2177]).

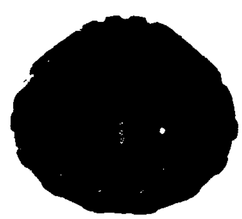

Fig. 4 Warm precision forging of an ultrahigh carbon steel (1.25% C) billet into a bevel gear. The forging temperature was 650°C.

stress exponents and microstructure, has led to the suggestion that superplastic flow may occur in limestones, $MgGeO_4$, mylonites (olivine-rich), and ice. It should be noted that at the extremely low strain rates of geological flow (as low as $10^{-14}s^{-1}$), superplastic mechanisms may operate at coarse grain sizes by comparison with conventional superplastic materials.

Strength of the Second Phase.

There is evidence to suggest that the relative strengths of the matrix and second phase constitute an important parameter in the control of cavitation during superplastic flow. Many fine-grained aluminum- and copper- base alloys are susceptible to cavitation during superplastic deformation and this is probably because of the large difference in strength between the matrix and the second hard phase. On the other hand, fine-grained Ti-6Al-4V and hypereutectoid $Fe-Fe_3C$ alloys do not exhibit cavitation. This may be attributed to the nearly similar strengths of the two phases making up the alloys at the temperature where superplastic flow is observed.

Size and Distribution of Second Phase.

If the second phase is considerably harder than the matrix phase it should be distributed uniformly and in fine particle form within the matrix. In the form of fine, but hard, particles, cavitation during superplastic flow is inhibited by various recovery mechanisms occurring in the vicinity of the particle. Chung and Cahoon (43) have shown how hard but fine silicon particles can minimize cavitation during superplastic flow of a fine-grained Al-Si eutectic alloy. Coarse particles, on the other hand, can lead to cavitation. For example, in white cast iron, cavitation occurs at the interface of coarse eutectic carbides ($\geq 10\mu m$) and the matrix during superplastic flow at 700°C. These large carbides cannot be refined by traditional thermal-mechanical working. It is possible, however, through rapid solidification technology and powder metallurgy, to obtain fine carbide particles distributed in the iron matrix in white cast irons. Ruano, Eiselstein, and Sherby (44) showed that a 3% C white cast iron produced in this way revealed virtually no cavitation after superplastic deformation, even after large tensile elongations in excess of 1000%.

Maehara (45) has a unique view on the influence of hard particles on enhancing superplasticity in stainless steels. He investigated a number of stainless steels and concluded that the optimal superplastic condition is achieved when a fine distribution of hard delta phase exists in the austenitic matrix. He obtained over 2000% elongation at 950°C and at a strain rate of $2 \times 10^{-3}\ s^{-1}$ for a stainless steel of composition 25% Cr, 7% Ni, 2.8% Mo, 0.85% Mr, 0.5% Si, 0.5% Cu, 0.3% W, and 0.14% N. Maehara considers that optimal superplastic behavior is achieved when recrystallization is occurring during deformation. He contends that recrystallization will only occur when a sufficiently large amount of hard, small particles exists (about 30% of sigma phase). The sigma phase particles act as sites for recrystallization because of severe heterogeneous deformation around the hard particles. Maehara concludes that soft particles distributed in a hard matrix will not exhibit superplastic behavior. His model certainly has considerable merit. It may not be applicable in other two phase systems, however, such as in white cast iron of eutectic and hypereutectic composition. Kim et al. (39) and Kum et al. (46) have shown that such a material, which consists of a continuous phase of relatively hard cementite, with a discontinuous soft phase of iron, is superplastic from 700 to 1025°C (see Fig. 6).

Nature of the Grain Boundary Structure.

The grain boundaries between adjacent matrix grains should be high angle (i.e. disordered). This is because grain boundary sliding (G.B.S.) is generally the predominant mode of deformation during superplastic flow. Low angle boundaries, as often obtained during warm working, do not readily slide under the appropriate shearing stresses. Structures containing low-angle grain boundaries in a eutectoid composition steel are not superplastic but can be made superplastic by converting the low angle boundaries to high angle ones by appropriate thermal or thermomechanical treatment (47). Similar results have been obtained in a tool steel (48) as well as an Al-Li based alloy (49).

Shape of Grains.

The grain shape should be equiaxed in order that the grain boundary can experience a shear stress allowing G.B.S. to occur. Materials with elongated cylindrical grains, even though fine grained in a transverse direction, would not be expected to exhibit very much grain boundary sliding when tested longitudinally. Testing in a transverse direction, however, could lead to extensive grain boundary sliding and therefore result in superplastic behavior.

Mobility of Grain Boundaries.

Grain boundaries in superplastic alloys should be mobile. During grain boundary sliding, stress concentrations develop at triple points, as well as at other obstructions along the grain boundary. The ability of grain boundaries to migrate during grain boundary sliding permits reduction of these stress concentrations. In this manner, grain boundary sliding can continue as the major deformation process. The fact that grains remain equiaxed after extensive superplastic deformation is indirect evidence that grain boundary migration is occurring. One possible explanation to the limited ductility observed in most fine grained ceramic polycrystals at high temperatures (even though the strain-rate-sensitivity exponent is high) is that the grain boundaries are not very mobile. In this case, the lack of boundary mobility can lead to high stress concentrations at triple points during grain boundary sliding, leading to crack nucleation and early failure.

Grain Boundaries and Their Resistance to Tensile Separation.

Grain boundaries in the matrix phase should not be prone to ready tensile separation. It is generally believed that the grain boundaries in ceramic materials have a high surface energy, and as a result will separate under low tensile stresses (50). This may be the major reason why many fine grained ceramic polycrystals exhibit low ductility in tension even when the strain-rate-sensitivity exponent is high. Perhaps the great success achieved in making polycrystalline zirconia superplastic (35-37) is because of the development of extraordinarily fine grains ($\bar{L}$=0.3μm) which reduced the stress required for plastic flow to a value below the grain boundary tensile fracture stress.

Optimizing the Rate of Superplastic Flow in FSS Materials

For a given superplastic material, and at a given temperature, there is a maximum strain rate where superplastic flow by grain-boundary sliding is no longer the dominant process; another mode of deformation becomes important, namely diffusion-controlled dislocation creep (slip creep). The maximum strain rate at which grain boundary sliding remains rate-controlling is typically on the order of 10^{-4} s^{-1}, a rate considerably lower than those used in most commercial forming operations (e.g., 10^{-1} to 1 s^{-1}). From a technological viewpoint it would be desirable to increase the maximum strain rate for superplastic flow. The approach for attainment of this goal is, in principle, straight forward. One must select structural variables that will enhance grain boundary sliding but make slip creep more difficult. We illustrate this schematically in Fig. 7(a). In this figure, the logarithm of the stress is plotted as a function of the logarithm of the strain rate. The two separate processes contributing to grain-boundary sliding and to slip creep are represented as straight lines. The point of intersection (marked $\dot{\varepsilon}_{SPmax}$) represents the maximum strain rate for superplastic flow for a given set of microstructural conditions. As noted on the figure, the principal equations for grain boundary sliding (governed by grain boundary diffusion) and for slip can be written as equations [1] and [2], respectively (9):

$$\dot{\varepsilon}_{gbs} = A_{gbs}\left(\frac{b}{\bar{L}}\right)^3\left(\frac{D_{gb}}{b^2}\right)\left(\frac{\sigma}{E}\right)^2 \qquad [1]$$

$$\dot{\varepsilon}_{slip} = A_s\left(\frac{\lambda}{b}\right)^3\left(\frac{D_L}{b^2}\right)\left(\frac{\sigma}{E}\right)^8 \qquad [2]$$

where $\dot{\varepsilon}_{gbs}$ and $\dot{\varepsilon}_{slip}$ are the strain rates for grain boundary sliding and slip, respectively, A_{gbs} and A_s are constants, b is Burgers' vector, $\bar{L}$ is the mean linear intercept grain size, λ is the minimum barrier spacing governing slip creep (typically the interparticle spacing or the grain size), D_{gb} is the grain boundary diffusivity, D_L is the lattice diffusivity, σ is the stress, and E is the dynamic, unrelaxed, Young's modulus. The point of intersection on Fig. 7(a) can be shown, by combining equations [1] and [2] and rearranging, to be:

$$\frac{\sigma}{E}_{\dot{\varepsilon}_{SPmax}} = \left[\frac{A_{gbs}}{A_s}\,\frac{D_{gb}}{D_L}\,\frac{b^6}{\bar{L}^3\lambda^3}\right]^{1/6} \qquad [3]$$

$$\dot{\varepsilon}_{SPmax} = \frac{(A_{gbs}D_{gb})^{4/3}}{(A_sD_L)^{1/3}}\cdot\frac{b^3}{\bar{L}^4\lambda} \qquad [4]$$

The generation of a new set of microstructural conditions to make grain-boundary sliding more facile and to inhibit the slip creep process, can increase the maximum strain rate for superplastic flow. Such a change leads to an increase in the maximum rate for superplastic flow from $\dot{\varepsilon}_{SPmax}$(i) to $\dot{\varepsilon}_{SPmax}$(ii) as indicated in Fig. 7(b) in which the dashed lines represent the new microstructural conditions. The most straightforward structural feature that can be modified to achieve such an enhancement in superplasticity is to decrease the grain size. When the grain size is reduced, the superplastic flow rate (by grain boundary sliding) is increased and the normal flow rate (by slip) is reduced (the Hall-Petch relation). In a further example, a bimodal distribution of second phase particles may lead to an ideal superplastic structure. The ultrafine particles can pin dislocation networks (subgrains) yielding a low value of λ, and fine particles can pin the fine grain size leading to a low value of $\bar{L}$.

A series of predictions are shown in Figs. 8(a), (b), and (c) from equations [1] and [2] for the case of a typical high stacking fault energy material. In Fig. 8(a), the log of the maximum strain rate for superplastic flow is shown as a function of the log of grain size, for a fixed value of λ (i.e., 1μm) with homologous temperature indicated as a variable. The figure demonstrates the significant influence of grain size on $\dot{\varepsilon}_{SPmax}$ at all temperatures. By a refinement of grain size from 100μm to a submicron value, the strain rate for superplastic flow can be increased from glacial flow rates to values typical of that for high rate forming operations such as extrusion. The influence of temperature is also significant; an increase from 0.55 to 0.72 Tm can improve $\dot{\varepsilon}_{SPmax}$ by an order of magnitude at a fixed grain size. It is important to recognize that the microstructure has to remain stable over this increased temperature range in order for these predictions to be realized.

In Figs. 8(b) and 8(c), the influence of the ease (or difficulty) of slip creep on the strain rate for superplastic flow is examined for the specific homologous temperatures of 0.60 and 0.72, respectively. In these figures, the log of $\dot{\varepsilon}_{SPmax}$ is plotted as a function of λ, the microstructural feature governing slip creep for a variety of grain sizes. As may be seen, the predominant influence on $\dot{\varepsilon}_{SPmax}$ is that of grain size; the role of λ, the microstructural feature that governs slip creep, is, however, also significant. For example, at a grain size of 2μm at

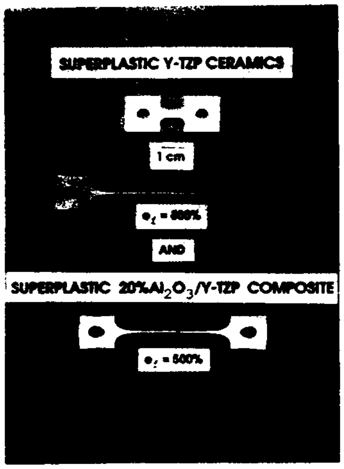

Fig. 5 Superplastic behavior in a yttria-stabilized, fine-grained zirconia, and a 20% Al_2O_3/Y-TZP composite ceramic (37,38).

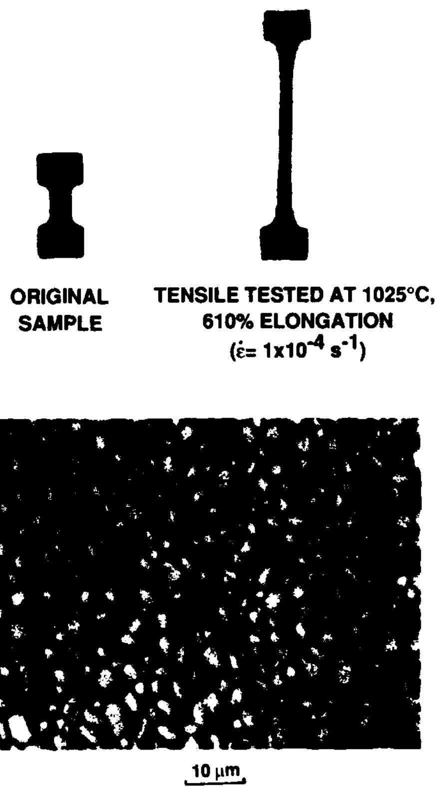

Fig. 6 Superplastic behavior in an iron carbide ceramic polycrystal containing 80% Fe_3C and 20% iron. The upper part of the figure shows a sample tested to 610% at 1025°C, and the photomicrograph shows the fine microstructure after superplastic deformation.

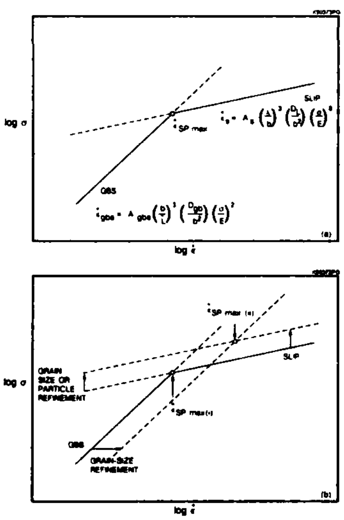

Fig. 7 The maximum strain rate for superplastic flow is represented by $\dot{\varepsilon}_{SPmax}$ and is the point of intersection on Fig. 7(a) between the equations for grain boundary sliding (gbs) and slip. By modifying the microstructural variables to increase the ease of grain boundary sliding and to inhibit slip, the maximum strain rate for superplastic flow can be increased from $\dot{\varepsilon}_{SPmax}(i)$ to $\dot{\varepsilon}_{SPmax}(ii)$ as indicated in Fig. 7(b).

800°C, refinement of λ from 1μm to 0.01μm is predicted to result in an improvement in $\dot{\epsilon}_{max}$ by two orders of magnitude from 10^{-3} to 10^{-1} s^{-1}.

A number of recent investigations have centered on aluminum alloys with the objective of increasing the strain rate range for superplastic flow by grain size refinement. These studies have shed some new insight into understanding the general behavior of superplastic materials and are discussed in the following subsections.

Whisker-Reinforced and Mechanically Alloyed Aluminum.

Wadsworth and his colleagues (51-54) have made considerable progress in achieving high elongations under high strain rate conditions. They have achieved this success by grain size refinement in whisker reinforced aluminum alloys and in mechanically alloyed aluminum alloys. In one example (51), a composite of Al-2124 containing 20 vol % SiC whiskers behaved in a superplastic-like manner (up to 300% elongation) at the high strain rate of 3×10^{-1} s^{-1}. In another example (52,53), an ultrafine grained, mechanically alloyed Al 9021 (4.2%Cu-2.0%Mg-1.1%C-0.8%O), exhibited maximum elongations at strain rates as high as 20 s^{-1}. In yet another investigation (54), on a similar mechanically alloyed aluminum alloy, MA-Al IN 90211, an optimum elongation of 500% was found at a strain rate of 2.5 s^{-1}.

An overview of the superplastic behavior of the alloys described above is given in Fig. 9 in which elongation to failure is plotted as a function of strain rate. The grain sizes for each class of alloy groups is indicated on the figure. Included in the figure are data for mechanically alloyed nickel-base alloys (MA 754 and IN 6000) by Gregory et al. (55) who noted high ductility at high strain rates in these fine grained materials. Wadsworth and his colleagues are currently investigating the precise, operative, deformation mechanisms in these alloys. The observation of superplastic-like behavior (m is slightly below the typical value of 0.5) at such high strain rates, however, is believed to be consistent with the extremely fine (submicron) grain size contained in these complex aluminum and nickel composites and alloys.

Al-Mg Base Alloys.

Recently, considerable effort has been devoted to superplasticity studies in Al-Mg alloys. Magnesium forms a solid solution in aluminum and up to 10% magnesium can be dissolved at high temperatures. Al-Mg solid solution alloys are known as Class I solid solutions and exhibit high strain rate sensitivity even when coarse grained (56). In the following, three separate studies are described on fine-grained Al-Mg alloys that exhibit conventional superplasticity.

Watanabe, Ohori, and Takeuchi (57) have modified an existing commercial Al-Mg alloy (AA 5083) by the addition of copper to make it superplastic. The alloy composition of AA 5083 is nominally 5%Mg, 0.7%Mn, 0.12%Fe, 0.13%Cr and 0.10%Si. They showed that, with progressive addition of copper (0.2, 0.4, and 0.6%Cu) the grain size was increasingly refined by a recrystallization treatment following hot and cold rolling. The grain size of the 0.6% Cu alloy was less than 10μm. The improvement in ductility with increase in copper content is illustrated in Fig. 10. An elongation of 700% was obtained at a strain rate of 2.8×10^{-3} s^{-1} at 550 °C, with a strain rate sensitivity exponent of 0.7, for the AA5083-0.6% Cu alloy. Typical mechanical properties at room temperature of superplastically formed parts from the new alloy are 334 MPa tensile strength and 25% elongation. These values are superior to those obtained in Supral alloy containing 5%Cu - 0.5%Zr in the superplastic condition (UTS - 181 MPa, 7% elongation). The newly-developed Japanese sheet has the trade name of NEOPRAL and is used for building and construction applications, mostly for decorative components.

Salama and McNelley (58) have been studying the properties of an Al-10% Mg alloy containing small additions of either zirconium (0.2%) or manganese (0.5%). These authors show that, when fine-grained, these alloys are superplastic at relatively low temperatures (300°C) in contrast to other superplastic aluminum alloys where superplasticity is only observed at high temperatures (475 to 550°C). In addition, the room temperature strength of the Al-10Mg alloys is superior to other Al-Mg alloys because of increased solid solution strengthening. Both alloys studied by Salama and McNelley behaved in a similar manner. The ductility characteristics and stress-strain rate relations of a fine and coarse grained Al-10Mg-0.2Zr alloy at 300°C are compared in Fig. 11. As can be seen, the coarse grained alloy ($\bar{L}$=40μm) exhibits a maximum m value of 0.33 with maximum elongations in the order of 200%. On the other hand, the fine grained alloy ($\bar{L}$=3μm) exhibits m values approaching 0.5 with a maximum elongation of 500%.

The superplastic properties of a powder metallurgy product based on Al-5Mg-1.2Cr have been studied by Shin, Selby, Belzunce, Roberts and Sherby (59). After consolidation of the powders by hot compaction, followed by thermal-mechanical processing, a fine-grained ($\bar{L}$=2μm) condition was achieved. The material exhibited high elongations (~1000%) at high strain rates ($\dot{\epsilon}$ =2 x 10^{-2} s^{-1}) when tested at high temperature (550°C). The creep behavior was similar to those observed by Salama and McNelley and the results are shown in Fig. 12. At intermediate strain rates, the fine-grained material exhibits a stress exponent of about two or less (m=≥0.5). At low strain rates, the stress exponent increases, and this increase occurs at reduced stresses as the temperature of testing is increased. A very significant observation is that the high stress exponent appears to decrease again at very low strain rates as the base-line for solute-drag-controlled dislocation creep is reached.

The trends shown in Fig. 12 suggest that there are two competing independent processes during deformation of fine-grained superplastic materials. These processes are grain boundary sliding with a threshold stress and dislocation-controlled slip. This concept is schematically illustrated in Fig. 13. The dash-dot line in the figure depicts the creep rate of a fine-grained material when grain boundary sliding with a threshold stress is the deformation process. The dashed line in the figure represents the creep rate of the material when dislocation creep

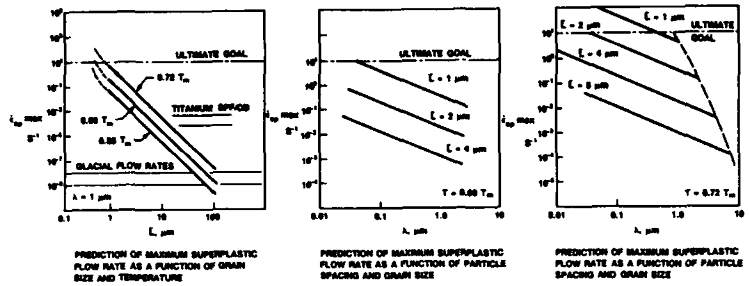

Fig. 8 In this figure, for the case of a typical high stacking fault energy material, the influences of temperature, grain size, and interparticle spacing, λ, on the maximum strain rate for superplasticity, $\dot{\epsilon}_{SPmax}$, are illustrated.

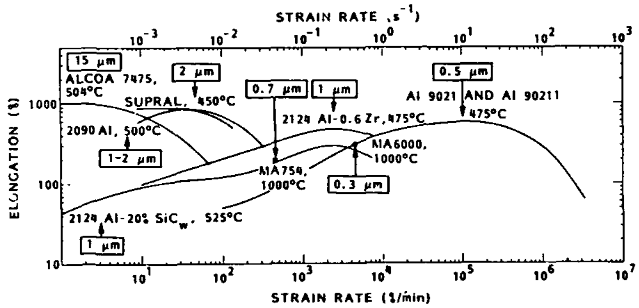

Fig. 9 Overview of superplastic behavior as a function of strain rate for a range of aluminum based materials of grain sizes from 15µm to sub-micron. Also included are data points for fine-grained nickel-based superalloys.

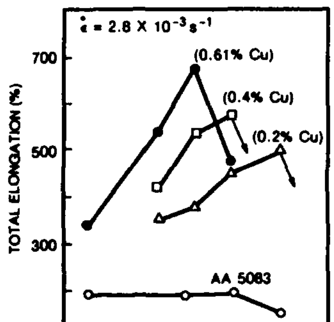

Fig. 10 Influence of copper additions on the tensile ductility of a fine-grained Al-5Mg alloy (AA 5083). Data from Wantanabe, Ohori and Takeuchi (57).

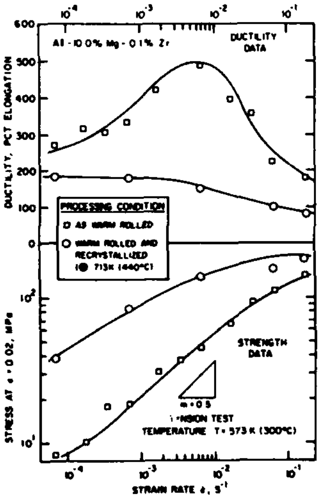

Fig. 11 Comparison of the ductility and flow stress-strain rate behavior of a fine- and a coarse-grained Al-10Mg-00.1 Zr alloy at 300°C. Data from Salama and McNelley (58).

in the matrix is the rate-controlling process. Since these two processes are considered to be independent processes, the fastest one is rate-controlling. The predicted behavior is given by the shaded curve. Four ranges are depicted: Ranges 0, I, II, III. Ranges 0 and III represent deformation controlled by slip, Range II is where superplastic flow is expected, and Range I is where a threshold stress is observed and a low ductility can be expected. The data in Fig. 12 indicate that the threshold stress is a function of temperature, decreasing in magnitude with increase in temperature. The actual meaning of threshold stress and its value as a function of temperature, as well as of grain size, is not clear and is an area of investigation worthy of considerable additional studies.

Mechanisms of Deformation in FSS Materials

It would appear that the structural prerequisites for fine structure superplasticity are quite well developed, and many new alloys have been developed in recent years with the specific objective of achieving superplastic characteristics. A detailed understanding of the exact mechanism of plastic flow in FSS materials, however, has not been as thoroughly developed. The most commonly-considered mechanism involves grain boundary sliding. It is generally thought that an accommodation process accompanies grain boundary sliding. This accommodation process might be grain boundary migration, recrystallization, diffusional flow or some dislocation slip process. Quantitative models have been developed to describe superplastic flow accommodated by slip recovery processes. Examples are those given by Ball and Hutchinson (60), by Mukherjee (61), and by Langdon (62). The "core and mantle" theory of Gifkins (63) also considers slip recovery mechanisms in the vicinity of the grain boundary. A different view is that of grain boundary sliding accommodated by diffusional flow, and such a model has been quantitatively developed by Ashby and Verrall (64). All of these models have some features that are in agreement with experimental observations in superplastic materials. The models, however, have not been able to predict quantitatively the creep rates actually observed in fine-grained superplastic materials. In addition, none of the theories are able to predict, in one relation, the correct stress, temperature and grain size dependencies (9).

Another obstacle to understanding mechanisms of superplastic flow is the existence of Range I creep (described earlier in association with Figs. 12 and 13). Whereas many researchers associate this low stress region with low strain-rate sensitivity, others have observed high strain-rate-sensitivity in this region, even in the same alloy system, and some fierce debates have taken place on this subject (65,66). Range I creep has been explained and discussed in various ways: for example, as a new deformation process controlled by slip (67), as a process associated with a threshold stress for plastic flow, and as a grain boundary sliding process with a reduced creep rate from grain growth (68). Another view, developed by Mayo (69), utilizes a modification of Gifkins' "core and mantle" model to explain Ranges I, II and III in the following way. The core is assumed to have a typical strain rate sensitivity exponent as is observed in coarse grained materials (i.e., $m=0.2$). The mantle region is considered to be weaker than the core region but otherwise has a similar strain-rate-sensitivity exponent. Mayo assumes that the mantle region, which has a finite width, increases in volume with a decrease in stress, and it is this change with stress that leads to the transition Range II. Such a model leads to values of m higher than 0.2 although a specific value of m is not predicted for Range II. Experimental work by Mayo on surface observations of grains in the superplastic region (involving Pb-Sn and Zn-Al alloys) suggests that the mantle region makes up a large fraction of the total material.

Typical strain-rate-sensitivity exponents observed in FSS materials are at values clustered around 0.5. This is especially true at intermediate temperatures and at fine grain sizes where the activation energy for superplastic flow is equal to that for grain boundary diffusion (9). At high temperatures, however, where the activation energy for superplastic flow is equal to that for lattice diffusion, the strain-rate-sensitivity exponent is equal either to 0.5 or to values greater than 0.5 (9). Fukuyo, Oyama, Tsai, and Sherby (70) have recently developed a model to explain the different results obtained at high temperatures. The model is similar to the Ball, Hutchinson, Mukherjee, Langdon concept based on a grain boundary sliding process accommodated by slip. The model is illustrated in Fig. 14. As can be seen, the slip accommodation process involves the sequential steps of glide and climb. When climb is the rate controlling step, the strain-rate-sensitivity exponent is 0.5 because of the pile-up stress at the head of the climbing dislocation (this, in fact, is the prediction of the Ball-Hutchinson, Mukherjee, Langdon relations). When glide is the rate-controlling step, however, the strain-rate-sensitivity exponent is equal to unity because there is no pile-up stress. Since glide and climb processes are sequential, the slowest of the two processes is rate controlling. This model predicts that fine-grained Class I solid solution alloys can exhibit a high value of m equal to unity since in these alloys the glide step (solute-drag-controlled dislocation creep) is often the slowest process (56,71). On the other hand, fine-grained Class II solid solution alloys, in which dislocation climb is the rate-controlling step, should exhibit m values that are not greater than 0.5.

The predictions of Fukuyo et al. have been confirmed for a number of fine-grained solid solution alloy systems studied at high temperatures, as can be seen in Figs. 15 and 16. In these figures, the strain-rate-sensitivity exponent is plotted as a function of the strain rate for fine-grained Class I solid solution alloys (Fig. 15) and for fine-grained Class II solid solution alloys (Fig. 16). The strain rate is normalized to the strain rate (in Range I) at which the strain-rate-sensitivity exponent, m, is equal to 0.3. In this manner, different materials can be assessed from a common base. The Class I solid solution alloys shown in Fig. 15 are Fe-10Al-1.25C (70), Ti-6Al-4V (72,73), Mg-33Al (74), Al-33Cu (75) and Al-25Cu-11Mg (76). The strain-rate-sensitivity exponent is about 0.5 at low strain rates, then increases with increasing strain rate to values as high as 0.8. This is the trend predicted from the Fukuyo et al. model wherein at low strain rates, the grain boundary sliding process is

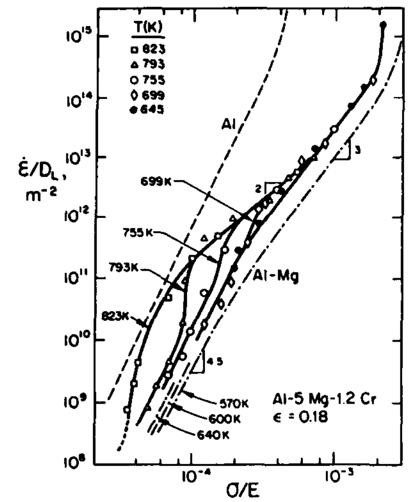

Fig. 12 Diffusion-compensated strain rate versus modulus-compensated stress for a PM processed Al-5Mg-1.5 Cr alloy ($\bar{L}$~2μm). High strain rate sensitivity is observed at intermediate stresses followed by a threshold stress region with decreasing stress. The threshold stress is a function of temperature.

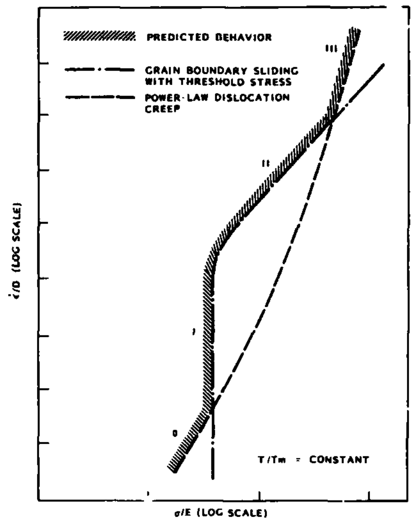

Fig. 13 Prediction of the creep behavior of a fine-grained material based on two independent processes: grain boundary sliding with a threshold stress and slip (power-law dislocation creep). Four ranges are depicted by 0, I, II and III.

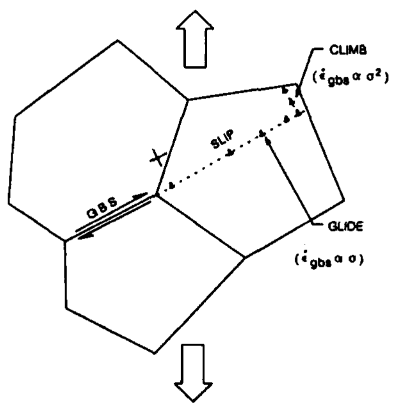

Fig. 14 Model illustrating grain boundary sliding accommodated by dislocation motion involving the sequential steps of glide and climb.

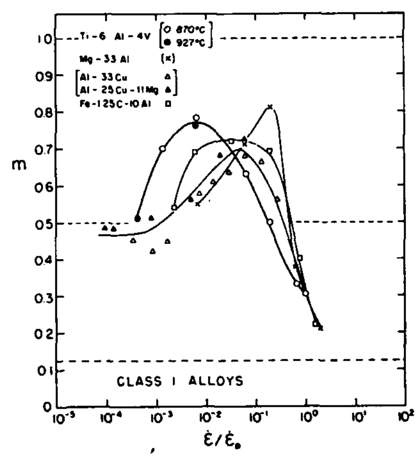

Fig. 15 Influence of normalized strain rate on the strain-rate-sensitivity exponent, m, for superplastic materials considered as fine-grained Class I solid solution alloys.

accommodated by climb, whereas at high strain rates it is accommodated by glide. These two regions are to be identified as part of Range II of Fig. 13. At yet higher strain rates, slip deformation becomes the dominant deformation mode and the m value decreases to low values in the order of 0.1 to 0.2. The superplastic materials shown in Fig. 16 exhibit trends that are quite different from those observed for the fine-grained Class I solid solution alloys. The four alloys listed are: Ni-39Cr-10-Fe-7.5Ti-1Al (77), Cu-39.4Zn (78), Fe-26Cr-6.5Ni-0.05Al (79), and Ag-28Cu (80). For these fine-grained solid solution materials, the strain-rate-sensitivity exponent is about 0.5 at low strain rates. With increasing strain rates, m decreases. This is the type of creep behavior expected in fine grained Class II solid solution alloys since the accommodation process, in this case, is grain boundary sliding controlled by dislocation climb where m equals 0.5.

SECTION B. INTERNAL STRESS SUPERPLASTICITY (ISS).

In addition to the fine structure superplasticity described in the previous section, there is another type of superplasticity known as internal stress superplasticity (10). In these materials, in which internal stresses can be developed, considerable tensile plasticity can take place under the application of a low, externally-applied, stress. This is because internal-stress superplastic materials can have a strain-rate sensitivity exponent of as high as unity, i.e., they can exhibit ideal Newtonian-viscous behavior. Such superplastic materials deform by a slip creep mechanism.

There are many ways in which internal stresses can be generated. These include thermal cycling of composite materials in which the constituents have different thermal expansivity coefficients (81-86), thermal cycling of polycrystalline pure metals or single phase alloys that have anisotropic thermal expansion coefficients (86,87), and thermal cycling through a phase change (88-92). It should also be noted that pressure induced phase changes have been cited as a source of superplastic flow in geological materials (42). For example, there is a transformation in the earth's upper mantle because of pressure from orthorhombic olivine to a spinel phase at a depth of about 400 km below the earth's surface. It is believed that internal stress superplasticity arising from transformation stresses through pressure cycling (analogous to temperature cycling) leads to a mixed phase region of low effective viscosity.

<u>Whisker and Particle Reinforced Composites.</u>

It has been shown that internal stress superplasticity can be utilized to enhance the ductility of metal-matrix composites that are normally brittle. Thus, a whisker-reinforced metal-matrix composite (aluminum containing 20%SiCw alloy) was made ideally superplastic, i.e., Newtonian viscous in nature, during deformation under thermal cycling conditions (81,82,84,85). An example of the exceptional tensile ductility that can be achieved in this manner, in a whisker-reinforced 6061 aluminum alloy, is shown in Fig. 17. Whereas the metal-matrix composite exhibits only 12% elongation under isothermal deformation at 450°C, the same composite exhibits 1400% elongation if deformed under thermal cycling conditions (100 $\leftrightarrows$450°C at 100 seconds per cycle). An example of similar behavior was demonstrated with a zinc-30 vol.% alumina particulate composite (82-83). Whereas this material exhibited essentially nil ductility when tested in tension at 300°C, it exhibited Newtonian-viscous behavior when deformed under thermal cycling conditions, and elongations exceeding 150% were achieved.

The basis of understanding the effect of internal stress on enhancing the ductility of metal-matrix composites is as follows (87). During thermal cycling, internal stresses are developed at the interfaces between the metal-matrix and the hard ceramic second phase. This is because the thermal expansion coefficient of the metal-matrix is several times larger than that of the ceramic phase. These internal stresses will relax by plastic deformation in the metal-matrix to the value of the local interfacial yield stress of the material. It is this remaining local yield stress, which we define as the internal stress, σ_i, which contributes to the low applied external stress, and results in macroscopic deformation along the direction of the applied stress. This model is developed quantitatively in the next subsection. The creep behavior of two 2024 Al-SiCw composites under both thermal cycling and isothermal conditions are shown in Fig. 18 (85). The graph shows a plot of the diffusion-compensated creep rate as a function of the modulus-compensated stress. Three trends can be noted. First, the thermally cycled composites are much weaker than the isothermally tested composites at low applied stresses. Second, the thermally cycled composites have strain-rate-sensitivity exponents of unity at low stresses. Third, the thermally cycled samples and the isothermally tested samples yield data that converge at high stresses; this is expected since the internal stress generated by thermal cycling (a constant) will have a diminishing contribution to creep as the applied stress is increased.

It was observed that the manner in which the silicon carbide whiskers flowed during plastic deformation of the composite was dramatically different under thermal cycling conditions by comparison with isothermal conditions (81,84,85). For example, Wu and Sherby (81) have shown that regions devoid of whiskers in a 2024-20 vol.% SiCw extruded composite become readily filled with whiskers when deformed in tension under thermal cycling conditions. Another example of the difference in the rate of reorientation of whiskers in the same extruded 2024-20 vol.% SiCw composite, was noted in tests conducted in compression under isothermal and thermal cycling conditions (85). This example is shown in Fig. 19. The sample that was deformed isothermally exhibited very limited reorientation of the whiskers towards the direction of compression flow. Furthermore, extensive surface cracks were observed in the sample. On the other hand, the sample that was deformed under thermal cycling conditions, to the same strain, showed that nearly all whiskers were no longer in the original longitudinal direction. Clearly, Newtonian-viscous flow of the metal-matrix accelerates the reorientation of the SiC whiskers. No surface cracks were observed in the thermally cycled sample.

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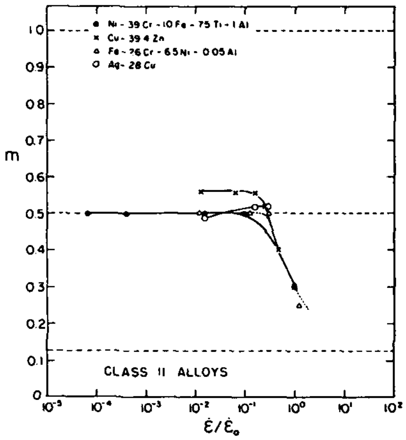

Fig. 16 Influence of normalized strain rate on the strain-rate-sensitivity exponent, m, for superplastic materials considered as fine-grained Class II solid solution alloys.

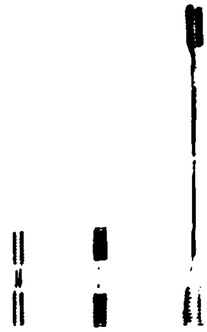

Fig. 17 Tensile ductility of 6061 Al-SiCw reinforced composite. The sample on the left hand side is in the untested condition. The center sample exhibits 12% elongation under isothermal testing at 450°C and at $\dot{\epsilon}=10^{-4}\ s^{-1}$. The sample on the right hand side exhibits 1400% elongation under thermal cycling conditions (100⇄450°C) at σ=10MPa. From G. Gonzalez et al. (84).

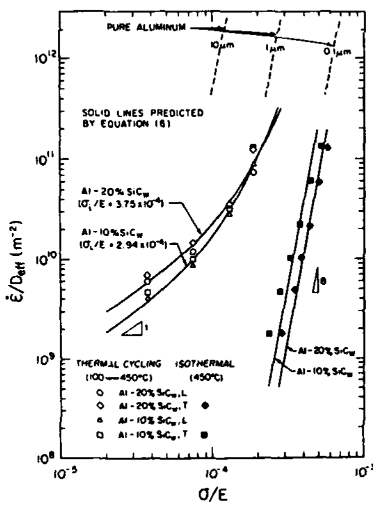

Fig. 18 Diffusion-compensated strain rate as a function of modulus-compensated stress for thermal cycled and isothermal creep data from 2024 Al alloy composite containing 10% SiCw and 20% SiCw.

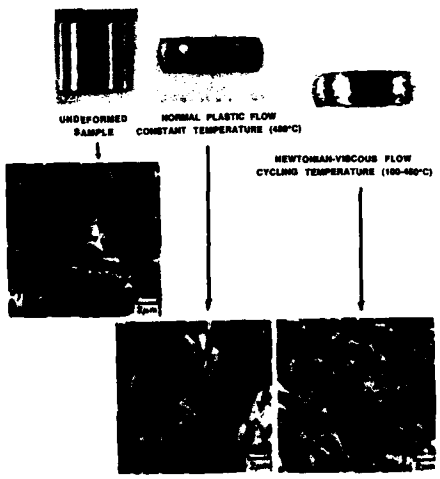

Fig. 19 The above figure illustrates that Newtonian-viscous flow in a whisker reinforced 2024 Al-20% SiCw composite under thermal cycling leads to crack-free plastic flow and to rapid reorientation of the whiskers.

Anisotropic Expanding Polycrystalline Materials.

Other examples of internal stress superplasticity are in polycrystalline zinc and alpha uranium. In the case of these two metals, the internal *stress arises from the anisotropy of* expansion coefficients present in these non-cubic structure materials. During thermal cycling internal stresses will be induced at the boundaries between adjacent grains.

A quantitative relation was developed by Wu, Wadsworth and Sherby to describe the thermal cycling behavior of anisotropic polycrystalline materials and composites (87). The model begins with the Garofalo hyperbolic sine creep relation (93), which describes the steady state creep rate of metals that are controlled by diffusion-assisted dislocation creep. This relation is given as:

$$\dot{\varepsilon} = (K/\alpha^n)D_{eff}/b^2[\sinh\alpha\sigma/E]^n \qquad [5]$$

where α and K are constants, b is Burgers' vector, E is Young's modulus, n is the stress exponent, D_{eff} is the effective diffusion coefficient and σ is the applied creep stress. The value of n is typically in the range 5 to 8.

The theory of Wu et al. (87) uses this equation on the assumption that the internal stresses generated during thermal cycling contribute to plastic flow in the following way: at any given time, half of the moving dislocations are aided by the presence of the internal stress, whereas the remaining half of the dislocations are opposed by the presence of the internal stress. The *theory also assumes that these two groups of dislocations contribute to plastic flow* independently of each other. The following equation results:

$$\dot{\varepsilon} = 1/2(K/\alpha^n)D_{eff}/b^2\{[\sinh\alpha(\sigma + \sigma_i)/E]^n + |\sigma - \sigma_i|/(\sigma - \sigma_i)[\sinh\alpha(\sigma - \sigma_i)/E]^n\} \qquad [6]$$

When the applied stress is much greater than the internal stress this expression leads to the Garofalo equation (eq. [5]). *In this case the prediction is that the material under thermal* cycling conditions approaches the isothermal behavior with a high stress exponent preventing attainment of superplastic properties. However, for the low applied stress range, for example on the order of (or lower than) the internal stress present within a material, equation [6] reduces to:

$$\dot{\varepsilon} = K\, n\, D_{eff}/b^2(\sigma_i/E)^{n-1}\sigma/E \qquad [7]$$

which is characterized by a stress exponent of n=1. This equation predicts that deformation in this range will result in Newtonian-viscous behavior and hence superplasticity is expected in the low applied stress range. This model correctly predicts the Newtonian flow behavior for a *number of materials under thermal cycling conditions (82,84-87). It was shown (87) that the* internal stress term can, in fact, be estimated without recourse to performing thermal-cycling tests. The value of the internal stress is given by

$$\sigma_i/E = [2.2 \times 10^{-4}\ b^2/t\, K\, D\,]^{1/n} \qquad [8]$$

where t is the cycle rate in seconds per cycle.

Equations [6] and [8] were utilized to predict the thermal cycling behavior of polycrystalline zinc and alpha uranium under different conditions of cycling and temperature ranges. The predicted curves are compared with experimental data in Fig. 20. As can be seen, the thermal *cycling data are well predicted by equation [6].* At lo. values of the modulus-compensated stress, the model predicts, and the data demonstrate, the exi. ence of ideal Newtonian-viscous flow (m=1); under these experimental conditions, superplastı. behavior is ind ed observed (86,87).

The thermal cycling data shown for the 2024-SiCw composite in Fig. 18 was also predicted from equation [6]. *As can be seen, the predicted curves fit very well with the experimental points.*

Materials Undergoing Polymorphic Changes.

Materials that undergo polymorphic changes with temperature can exhibit Newtonian-viscous *behavior when tested under thermal cycling conditions.* Among the early investigations on this subject were those of Sauveur (88), Wasserman (94), de Jong and Rathenau (89), and Clinard and Sherby (90) on iron base alloys. Oelschlagel and Weiss (95) showed that large elongations can be achieved in this manner. The internal stress arising in this case is from the difference in volume between the two phases during phase transformation. Such behavior is sometimes known as transformation plasticity. Application of ductility enhancement by cyclic phase transformation has been attempted with polymorphic ceramics. Many of these ceramics are based on bismuth oxides. It was shown (96-99) that high strain-rate-sensitivity exponents were obtained in many of these ceramic-base materials under thermal cycling conditions. Wadsworth and Sherby (92) have reviewed these studies and have concluded, however, that none of these ceramic systems have been shown to be superplastic. All tests were done in compression. Because tensile tests were not performed, measurements of high values of the strain-rate-sensitivity exponent, m, in compression are inconclusive evidence for superplasticity. *This is a consequence of the fact* that high values of m are a necessary but insufficient criterion for superplasticity. Since polycrystalline ceramics are susceptible to grain boundary separation, it is possible that, in the ceramic oxides studied *for transformation plasticity, they may well have failed with* negligible ductility in tension. Clearly this uncertainty should be clarified with appropriate additional experimental studies.

Interest in phase transformation plasticity appears to have been revived recently. Notable are the studies of Tozaki, Uesugi, Okada and Tamura on iron-base alloys (100), and of Furushiro, Kuramoto, Takayama and Hori on commercially pure titanium (101). Saotome and Iguchi (102) made in-situ microstructural observations during transformation superplasticity in pure iron, and concluded that interphase boundary sliding was an important deformation mechanism. Such studies are worthy of renewed pursuit, especially with respect to materials that are normally difficult to fabricate (such as polymorphic ceramics and intermetallic compounds). The ISS model presented earlier for describing the behavior of thermally cycled anisotropic polycrystals and metal-matrix composites may well be applicable to describing phase transformation plasticity.

SECTION C. ENHANCED POWDER CONSOLIDATION THROUGH SUPERPLASTIC FLOW

A practical method for superplastic forming of bulk material is through the use of powder metallurgy methods. The approach here is to achieve net-shaped products, with high density, by compaction of powders utilizing fine structure or internal stress superplasticity methods. Studies have been performed by Ruano et al. (91) on the use of internal stress superplasticity in enhancing the densification of white cast iron powders. Caligiuri (103), as well as Isonishi and Tokizane (104) have utilized fine structure superplasticity to enhance the densification of ultrahigh carbon steel powders.

ISS Compaction of White Cast Iron Powders

The advent of new technologies centered on rapidly solidified powders often requires development of methods of enhancing densification wherein the fine structures present in such powders are retained. To achieve this goal it is necessary to use low temperatures, but this usually requires the application of high pressures if a high density is to be achieved. High pressures are often a limiting factor in the manufacture of powder products. One method of enhancing densification of powders is by accelerating plastic flow through the generation of internal stresses during warm pressing. As previously described, one technique for generating internal stress is through the use of multiple, solid-state phase transformations.

Ruano et al. (91) showed that the densification of rapidly solidified white cast iron powders was enhanced by multiple phase transformations. The basis of this result is that, during phase transformation volume changes occur which create internal stresses. These internal stresses assist plastic flow, and result in enhanced pressure-sintering kinetics. An example of such a result is shown in Fig. 21. The densification kinetics of white cast irons are shown as a function of applied stress under both isothermal and thermal cycling conditions. The results demonstrate that thermal cycling, under stress, is an important factor in enhancing densification. For example, under a very low externally-applied stress of 6.9 MPa, a density of 95% is found for ten cycles, and a density of 90% is found for one cycle. By contrast, densities of much less than 80% are found for the isothermally warm-pressed samples at both 650°C and 775°C. Transformation cycling is also seen to enhance the densification of white cast iron powders at high applied stresses. For example, at 20 MPa, a density of about 99% is found for 10 cycles and about 95% for one cycle. Without transformation cycling, a density of only 90.5% is found after warm pressing for 0.5 hour at 775°C and less than 80% is found after warm pressing for 0.5 hour at 650°C. The results shown in Fig. 21 indicate that high densification can be achieved in a short time by utilizing transformation cycling under small applied stresses.

FSS Compaction of Ultrahigh Carbon Steel Powders

Caligiuri (103) studied the pressure-sintering kinetics of metal powders. Specifically, the kinetics of pressure sintering of ultrahigh carbon steel (1.6%C) powders containing either coarse-grained (non-superplastic) structures or fine-grained (superplastic) structures were investigated. For comparison, the pressure sintering kinetics of iron powders were also investigated. In order to interpret the densification mechanisms of powders under an applied stress, the creep behavior of the ultrahigh carbon steels containing coarse and fine microstructures was established. The creep rate of the fine-grained steel at a given stress was higher than the creep rate of the coarse-grained steel. The densification of coarse- and fine-grained UHCS powders was found to parallel their creep behavior. The fine-grained powders densified more readily than the coarse-grained powders. From these results, it was concluded that a superplastic microstructure enhances densification and permits the pressing of powders into high density compacts at intermediate temperatures and low pressures. From the densification and creep studies, Caligiuri developed an equation which showed that the intermediate-stage densification rate ($\dot{\rho}$) can be related to the steady-state creep rate ($\dot{\varepsilon}_{ss}$) at a given temperature by:

$$\dot{\rho} = A\,[\,(1-\rho)/\rho]^n \;\; \dot{\varepsilon}_{ss} \qquad [9]$$

where n is the stress exponent and A is a constant (of value 57). Thus, for a fixed value of relative density, the densification rate at a given intermediate temperature and low pressure is directly proportional to the steady state creep rate of the material (under the same conditions of testing) and to the stress exponent of the material. Through eq. [9], the densification behavior of a material can be predicted by simply knowing its creep properties at the temperature and pressure of interest.

Since superplastic materials generally creep faster than non-superplastic materials, at intermediate temperatures and low stresses, eq. [9] predicts that superplastic materials will densify faster than non-superplastic materials during pressing. Furthermore, the relationship between densification rate and creep rate predicts that, through the term $((1-\rho)/\rho)^n$, superplastic materials (low stress exponent) will densify faster than non-superplastic materials (high stress exponent) at the same creep rate. This prediction was verified experimentally. First, creep experiments were performed to determine the stress, at 650°C, at which a superplastic 1.6%C steel has the same steady state creep rate as a non-superplastic commercially pure iron. [This stress was determined to be 43.1 MPa at 650°C.] Powders of these materials were then compacted at 43.1 MPa. The results of these experiments are shown in Fig. 22. The 1.6%C superplastic powders densify more rapidly than the iron powders.

SECTION D. SUPERPLASTIC LAMINATED COMPOSITES

An important property of fine-grained superplastic materials is that they are often readily bonded in the solid state, either to themselves or to other non-superplastic materials. This bonding is possible because of the presence of many grain boundaries in the fine-grained material which act as short-circuit paths for diffusion (105). The same materials, if coarse-grained, will not readily bond under identical conditions of pressure, temperature, and time.

The ease of solid-state bonding in fine-grained UHC steels makes it possible to prepare ferrous laminated composites with sharp interfaces between layers. Discrete interlayer boundaries are achieved because the laminated composites are prepared by roll bonding at low temperatures (e.g., 650°C). Such laminated composites have been shown to exhibit unique impact and superplastic properties. For example, very low ductile-to-brittle transition temperatures (-140°C) are obtained in a UHC steel/mild steel laminated composite in Charpy V-notch impact tests (106). This is attributed to notch blunting of the crack by delamination at the layer interfaces. Another useful characteristic of the UHC steel laminated composite is its intermediate temperature ductility properties. Thus, it is possible to make non-superplastic mild steel behave in a superplastic-like manner at intermediate temperatures by lamination to superplastic UHC steel (107). Strain-rate sensitivity exponents over 0.30 and elongations to fracture of over 400% were obtained. The strain rate-stress results show good agreement with constitutive equations for creep based on an isostrain deformation model. This model was used to predict the conditions of strain rate, temperature, and the percentage of non-superplastic component required to achieve nearly ideal superplasticity in a ferritic stainless steel clad to a UHC steel. Daehn, Kum, and Sherby (108) have shown that the predicted conditions are achieved at 825°C and at $\dot{\varepsilon}=10^{-3}\ s^{-1}$ for a UHC steel clad with a ferritic stainless steel (12% by volume). This combination of components and test conditions leads to the unexpected result that coarse-grained stainless steels can be made superplastic as shown in Fig. 23. Daehn (109) has performed gas pressure blow forming experiments with stainless steel and with stainless-steel-clad ultrahigh carbon steel to assess die-filling capabilities. The experimental arrangement and typical results are shown in Fig. 24. The left hand side photograph shows a stainless steel plate formed by gas pressure at 775°C. The sample ruptured after five minutes without properly filling the die; the radius of curvature at the apex is 16 mm. The right hand side photograph illustrates a stainless steel/UHCS laminate part formed by the same process and under the same test conditions as the stainless steel. In this case, the sample did not rupture and filled the die satisfactorily.

SECTION E. OTHER POSSIBLE SUPERPLASTICITY MECHANISMS

In addition to the detailed descriptions of superplastic behavior given in the previous sections, there are other examples of superplastic or potential superplastic mechanisms, that will be discussed briefly.

Class I Superplasticity

Class I solid solutions are a group of dilute alloys in which the glide segment of the glide/climb dislocation creep process is rate controlling because solute atoms impede dislocation motion (56,70,71,110). This group of alloys is of interest in superplastic studies because, as a result of the glide-controlled creep mechanism, they have an intrinsically high strain rate sensitivity of about m=0.33 (over certain temperature and strain rate ranges). The intrinsic nature of the high strain rate sensitivity is important because it suggests that complex thermomechanical processing, such as that needed for fine-grained superplastic alloys, is unnecessary in Class I solid solutions.

Because the strain rate sensitivity in Class I solid solutions is not as high as that found in classical superplastic materials (i.e., m=0.33 versus m=0.5 to 1.0), the elongations to failure are typically more modest, i.e., about 200 to 400%. It is probable that early reports of superplasticity in coarse-grained alloys are, in fact, the result of Class I solid solution behavior.

Alloys are listed in Table I that exhibit a strain rate sensitivity of m~0.33 and that are also believed to fulfill the criteria for Class I solid solutions. These criteria include atomic size mismatch and modulus (56,71) as well as chemical diffusion considerations (110). It will be noted in Table I that coarse-grained Class I alloys based on W, Nb, and Al all demonstrate large tensile elongations at elevated temperatures (T/Tm>0.4). An example of creep behavior exhibiting a value of n=3 (m=0.33) with concurrent, uniform, large, tensile elongation is shown in Fig. 25 for a Nb-10Hf-1Ti alloy (112). It should also be noted that Yaney and Nix (121) have pointed out that the addition of solute atoms is not the only way to cause dislocations to move in a viscous manner. Lattice friction effects may also reduce the glide mobility. This may be important in covalently-bonded solids (such as Ge and Si) as well as in ordered intermetallic compounds. In this latter group, strong repulsive forces exist between atoms of like character. Thus, the deformation behavior of B2 compounds (e.g. CoAl, NiBe,

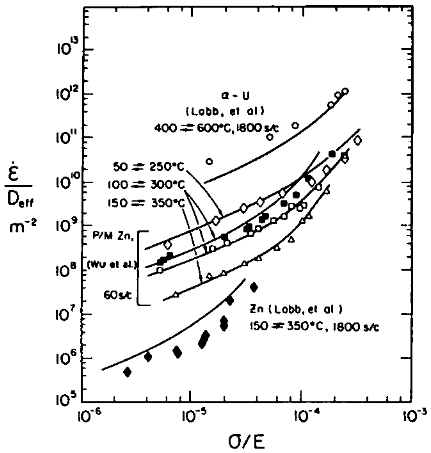

Fig. 20 Comparison of the internal stress creep model (Eqs. [6] and [8]) with all thermal cycling data available for zinc and alpha uranium.

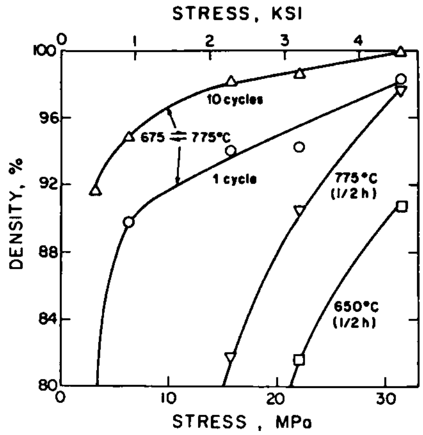

Fig. 21 Influence of multiple phase transformations (internal stress superplasticity) on the densification of white cast iron powders as a function of stress. High densities are achieved at low applied stresses under thermal cycling conditions.

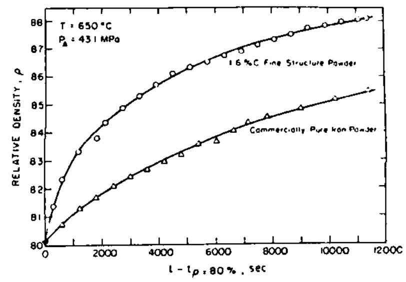

Fig. 22 Relative density-time curves for fine-grained 1.6%C UHC steel powders and commercially-pure iron powders compacted at 650°C. A constant applied pressure of 43.1 MPa was used, at which stress the creep rates of the two dissimilar materials are identical. The data is for the intermediate stage of compaction.

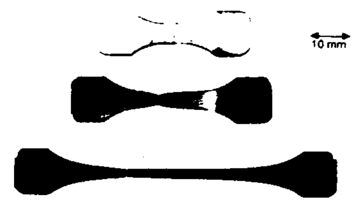

Fig. 23 Ferritic stainless steel samples in three conditions. The top specimen is in the untested condition, the center specimen is stainless steel tested at 825°C and $\dot{\epsilon}=10^{-3}$ s^{-1}, and the bottom spe[illegible]men is stainless steel clad to fine-grained ultrahigh carbon steel tested at 825°C and $\dot{\epsilon}=10^{-3}$ s^{-1}.

Stainless steel part formed by gas pressure. Sample ruptured without properly filling the die.

Stainless steel UHCS laminate part formed by gas pressure at 775 C. The sample did not fail and filled the die properly.

Fig. 24 Forming capabilities of cones for the cases of (LHS) stainless steel and (RHS) superplastic laminated stainless steel/UHCS.

TABLE I

Alloys Exhibiting Class I Solid Solution Behavior

ALLOYS	m	$\bar{L}$,µm	%El(max)	Ref.
W-33 at % Re	0.2-0.3	50-400	260	(111)
Nb-10Hf-1Ti	0.33	75	>125	(112)
Nb-5V-1.25Zr	0.26	ASTM #8	170	(113)
Nb-29Ta-8W-0.65Zr-0.32C	0.28	NA*	>80	(114)
Al-5456(Al-5Mg base)	0.33	20	153	(115)
Al-2 to 4% Ge	0.3-0.6	100-200	200-260	(116)
Al-Ca-Zn	0.25	2	NA*	(117)
Pu (Ag, Al, Zn impurities)	0.33	100	680	(118)
Mg-5.5 Zn - 0.5Zr	0.33	NA*	~100	(119)

Other materials showing m=0.33, (56,70,71,110,120)

In-Pb	Pb-In	Au-Ni	NaCl-KCl	Fe-Al-C
In-Hg	Pb-Sn	Ni-Sn	Mg-Al	Al-Cu-Mg
In-Sn	Pb-Cd	β-brass	Ti-Al-V	
In-Cd	Cu-Sn	Ni-Au	Al-Cu	

*** NA - Not available**

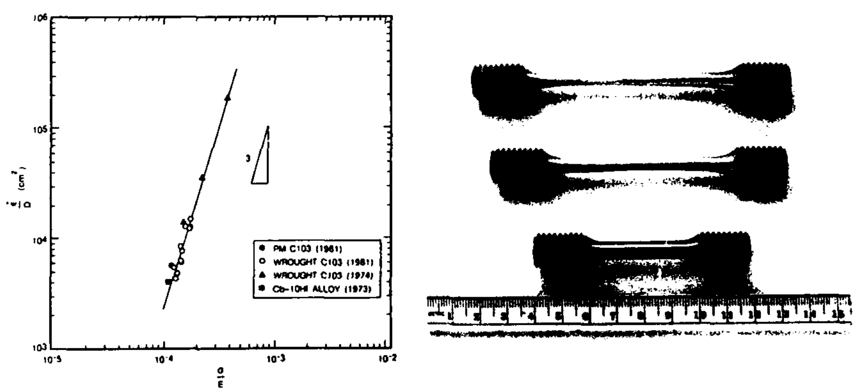

Fig. 25 Class I solid solution behavior in a C103 (Nb-10Hf-1Ti) alloy over the temperature range 1400-1700°C from several studies exhibiting n=3(m=0.33) (left-hand side) and concurrent large, uniform, tensile plastic strains (over 100%) (right-hand side) (104).

NiAl), in which lattice friction effects limit glide mobility may be quite similar to that of Class I solid solutions.

Two other examples in Table I, for nominally pure Pu and for a Mg alloy, also probably exhibit this type of behavior but are systems that require further experimental data. Other Class I solid solutions are given in Table I; in these cases, no measurements of tensile elongations have yet been made.

Viscous Creep Mechanisms for Superplasticity

As described previously, a high strain rate sensitivity is a necessary, albeit insufficient, prerequisite for superplastic flow. There are at least three creep mechanisms that yield a stress exponent, n, of unity and therefore behave in a *newtonian viscous manner* (i.e., m=1). These mechanisms are the two diffusional-based mechanisms of Nabarro-Herring (N-H) and Coble creep, and the slip-creep mechanism of Harper - Dorn (H-D) creep. It should be noted that Coble creep has often been considered and incorporated into models of superplastic flow (12). Furthermore, because recent studies (122-124) suggest that H-D creep usually predominates over N-H creep, the potential for H-D creep to yield superplastic behavior will be emphasized.

H-D creep is a slip-creep mechanism that has been the subject of increased research in recent years (122-124). The mechanism yields a stress exponent of unity and is usually observed at high temperatures and modest strain rates. A deformation mechanism map is shown in Fig. 26 in which grain size compensated by Burgers vector is plotted on the abscissa and modulus compensated stress is plotted on the ordinate at a fixed homologous temperature. It is noted that H-D creep dominates at relatively coarse grain sizes and low stresses. The metals shown on this map were formerly believed to deform by N-H creep; at that time, H-D creep was not considered as an alternative mechanism.

Because H-D creep has, to the present time, only been observed at low stresses, the associated low creep rates have not permitted measurements of strain to failure. The fact that ideal Newtonian viscous behavior is found in materials deforming by the H-D creep mechanism leads to the intriguing possibility of extensive plastic flow in these pure metals and alloys. This could be of technological importance if H-D creep can be developed at commercially - interesting strain rates, i.e., $\dot{\varepsilon} > 10^{-4}\ s^{-1}$. The likelihood of such a development depends upon an understanding of the physical basis of H-D creep. It is currently believed that the origin of H-D creep is the same as that obtained in ISS materials (described in Section B). The internal stress in the case of H-D creep arises from the presence of random, stationary, dislocations existing within subgrains. The equation for H-D creep is similar to Eq. [7]. The key physical parameters controlling the strain rate at which H-D creep can be expected to dominate include high dislocation density, low stacking fault energy, and high self diffusivity. High dislocation densities can be generated by testing in a radiation or ultrasound environment. Studies for achieving H-D creep under these special environmental conditions appear to be worth further pursuit to achieve superplasticity at technologically useful strain rates.

Ultrahigh Strain Rate Superplasticity

The final example of a viscous-like creep mechanism is that found at extremely high strain rates, i.e., up to strain rates of approximately 10^3 to $10^5\ s^{-1}$. Such high strain rates are in excess of those used in high rate forming operations such as explosive forming and rapid omnidirectional compaction. The evidence for possible superplastic behavior at strain rates of 10^4 to $10^5\ s^{-1}$ has been previously reported (125). In this work, flash X-rays were taken of Al and Cu shape charged liners undergoing large apparent tensile elongations as shown in Fig. 27. In these experiments, a shape charged liner (i.e., a hollow cone) undergoes a shape change to a long rod, in the solid state, at extremely high strain rates of up to $10^5 s^{-1}$. The x-ray data support the contention that deformation is occurring in the solid state, i.e., well below the liquidus of the metals.

It should be noted that there are a number of studies of deformation at high strain rates in which high strain rate sensitivites have been reported, for example, in LiF crystals (126). An example of raw data on metals from recent work by Follansbee (127) is shown in Fig. 28; as may be seen, apparently-high strain rate sensitivites are found at ultrahigh strain rates. The interpretation of such data, however, has been discussed by Follansbee who has questioned whether such high strain rate sensitivites are real, or are instead the result of testing techniques as well as structural changes with strain rate. Kunze and Mayer (128) have found similar trends in strain rate sensitivity increases with increase in strain rate at very high strain rates for a number of commercial steel and titanium alloys. They have also evaluated the tensile ductility of a number of their alloys at high strain rates and have determined that the elongation increases with increase in strain rate. For example, for a Ti-6Al-4V alloy tested at room temperature, their data reveal the following relation between m and % elongation-to-failure: at 2 s^{-1}, m = 0.01 and e = 14%, at 2 x $10^2\ s^{-1}$, m = 0.025 and e_f = 16%, and at 2 x $10^3\ s^{-1}$, m = 0.095 and e_f =18%. Extrapolation of these results to strain rates of 10^4 to $10^5\ s^{-1}$ would suggest the possibility of achieving high strain rate sensitivities and associated high ductilities.

It is worth pointing out that in the case of explosive forming, high strain hardening rates (as opposed to high strain rate sensitivity) is often used to explain the capability of metals to undergo extensive plastic deformation. In the case of the plasticity described for the example of shape charged liners this viewpoint should also be considered. It is important to note, however, that the phenomenon is observed not only in the FCC metals of Al and Cu, but also in the BCC metals Ta and W. Unlike FCC metals, the BCC metals show very limited strain hardening characteristics as strain rates increase, and it is unlikely that such effects can account for

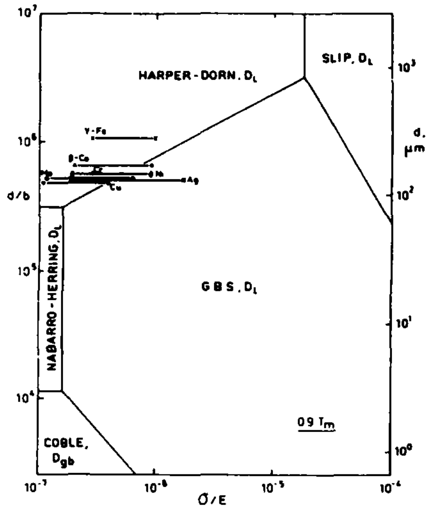

Fig. 26 Deformation map for metals and dilute alloys at a high homologous temperature ($0.9T_m$). H-D creep dominates at low stresses and coarse grain sizes and has been used to explain the Newtonian-viscous creep behavior of the pure metals shown in the figure (123).

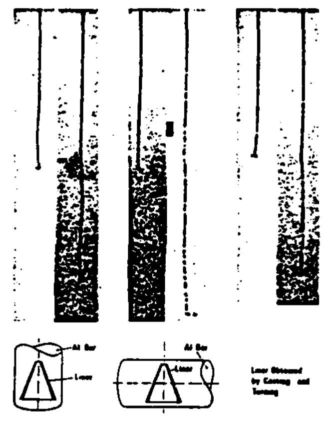

Fig. 27 Flash radiographs of an aluminum rod undergoing large strain plastic deformation at extremely high strain rates (125).

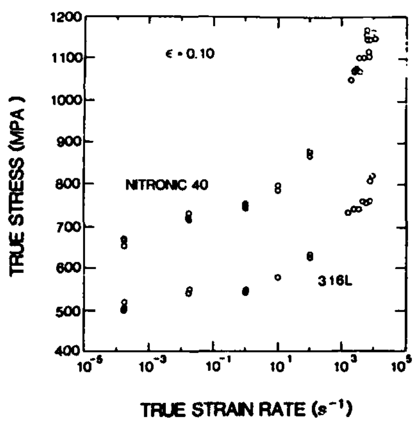

Fig. 28 Apparent Newtonian viscous flow is observed at very high strain rates in two FCC alloys (127).

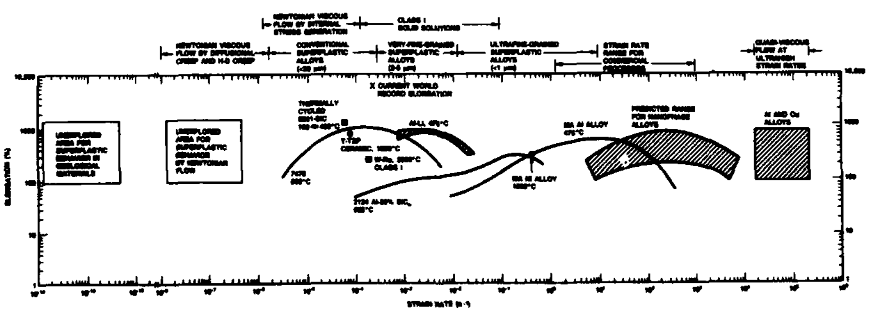

Fig. 29 Overview of documented superplastic behavior as a function of strain rate, including possible areas in which superplastic behavior may be observed. (Note the break in the strain rate scale at 10^{-8} s^{-1}).

the extremely high plastic deformation that is observed. The area of potential superplastic flow at such high strain rates is an intriguing one. To place the phenomenon in perspective, an overview of the various types of superplastic behavior described in this review, in the form of superplastic elongation as a function of strain rate, is presented in Fig. 29.

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CAVITATION AND SUPERPLASTICITY

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SUMMARY

Cavitation occurs in many alloy systems during superplastic flow. Cavities either pre-exist or nucleate on grain boundaries and their subsequent growth, coalescence and interlinkage leads to premature failure. The presence of cavities in superplastically formed components to be used for load bearing applications is clearly undesirable. It is apparent from studies on a wide range of materials that cavity growth is dominated by matrix plastic flow and that coalescence plays an important role in the development of large cavities. Hence, if cavitation damage is to be prevented, it is necessary to inhibit the nucleation event and to avoid the presence of pre-existing defects by careful control of the processing required to produce the superplastic microstructure. The influence that microstructural features and deformation conditions have on cavity nucleation is examined for a number of alloy systems, and it is clear that it is difficult to control these parameters so as to totally avoid cavitation. However, cavitation can be eliminated by the application of a hydrostatic pressure during forming. This reverses the sense of the driving force for cavity growth and the conditions for zero growth, which depend on the geometry of deformation, have been identified. Cavitation damage can be prevented by superimposed pressures of 0.5 to 0.75 of the uniaxial flow stress, although lower levels of pressure can substantially reduce the extent of cavitation.

1. INTRODUCTION

Despite the large plastic strains that may be achieved it is well established that cavitation may occur during superplastic flow [1-3]. Alloys for which cavitation has been reported include those based on aluminium [4-8], copper [9-15], iron [16-19], lead [20], magnesium [21], silver [22], titanium [23-25] and zinc [26-29]. The cavities nucleate on grain boundaries as a result of incomplete accommodation of grain boundary sliding, and their subsequent growth, coalescence and interlinkage leads to premature fracture. There is also evidence that cavities may grow from pre-existing defects such as cracks, associated with hard intermetallic particles or inclusions, which form during the thermomechanical processing required to develop superplastic microstructures [30].

In non-cavitating alloys failure during superplastic tensile deformation occurs as a result of unstable plastic flow and usually results in the material pulling down to an almost negligible cross-sectional area at fracture, as seen in the upper specimen in Fig. 1 for uniaxial straining. On the other hand, tensile failure resulting from the interlinkage of cavities leads to a macroscopically flat fracture face as seen in the lower specimen in Fig. 1.

Figure 1. Fractured tensile specimens showing failure by unstable plastic flow (Ti-6Al-4V) (upper), and cavitation failure (7575 Al alloy) (lower). Elongation ~ 900%.

Apart from the effect of cavitation on failure mode, a much more important consideration is its influence on the service behaviour of superplastically formed parts. As many load bearing components are being, or will be, used in the aerospace industry, the presence of cavitation damage and its elimination will be an important consideration in the development of superplastic materials and forming processes. A significant requirement for cavitation is the presence of a local tensile stress. Under conditions of homogeneous compression cavitation is not observed and cavities which have been produced during superplastic tensile flow are removed during subsequent compressive flow [31,32]. Superplastic closed die forging of nickel-base superalloys such as IN100, starting from hot pressed powders which are heavily worked by extrusion, has been used in the manufacture of turbine discs to give a sound cavity free product of uniform microstructure [33]. It has been demonstrated that the superimposition of hydrostatic pressure during uniaxial and biaxial superplastic tensile deformation can reduce or eliminate cavitation [34]. Cavitation can also be removed by a post-forming hot isostatic pressing treatment [35].

To control cavitation it is important to understand the microstructural and deformation parameters which influence its occurrence. The present paper examines the factors which influence cavity nucleation, the processes which may be involved in cavity growth, and the role of cavity coalescence. Failure in cavitating systems is also outlined. The ways in which cavitation damage may be minimised are considered and, in particular, the role that imposed hydrostatic pressures have on cavitation during superplastic flow is discussed.

2. CAVITY NUCLEATION

Strain is accumulated during superplastic flow primarily as a result of grain boundary sliding. Deformation occurs topologically and the formation of voids becomes geometrically necessary if rigid grains are to move relative to each other [36,37] Fig. 3. Fortunately, however, the relative displacements can be accommodated by the redistribution of matter within a narrow zone, or mantle, adjacent to the grain boundary of width ~0.07d, where d is the grain diameter [38]. This accommodation can be achieved by grain boundary and volume diffusion, the movement of grain boundary dislocations, or by the glide and climb of dislocations across the grains. The driving force for these processes is provided by local variations in the magnitude of the unrelaxed grain boundary normal tractions and shear stresses [40-42]. When the accommodating processes fail to meet the requirements imposed by the deformation rate then the stresses are not relaxed sufficiently quickly and cavities nucleate.

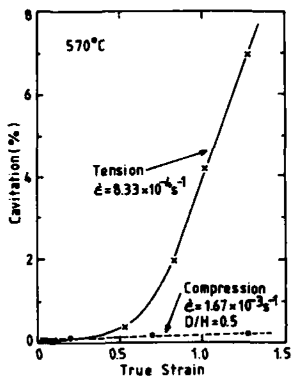


Figure 2. Cavitation behaviour of a superplastic α/β Cu-Ni-Zn alloy deformed in tension and compression.

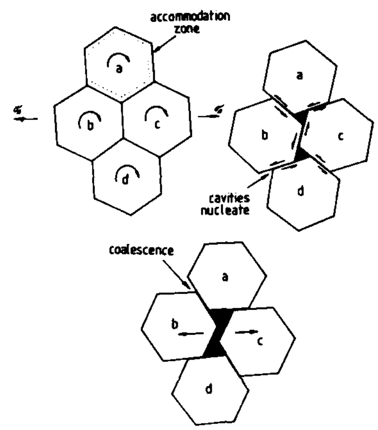


Figure 3. Representation of void nucleation, growth and coalescence, due to unaccommodated grain boundary sliding.

The factors which influence cavity nucleation include those which relate to microstructure such as grain size, the type, size, volume fraction and distribution of hard particles, the proportions and physical properties of the major phases, and those which are associated with the deformation conditions such as strain, strain rate, temperature and stress state. As will be seen there is often a strong interaction between microstructure and deformation conditions.

Stowell [30] has proposed a relationship from which the critical strain rate, $\dot{e}_c$, below which cavity nucleation at a grain boundary particle of diameter, D, is likely to be inhibited by diffusional stress relaxation can be predicted. This is given by:

$$\dot{e}_c = \frac{11.5\sigma\Omega}{xdD^2} \cdot \frac{D_{gb}}{kT} \quad (1)$$

where d is the grain diameter, x is the fraction of the total strain carried by grain boundary sliding, D_{gb} is the grain boundary diffusion coefficient, σ, the flow stress at the imposed strain rate, and Ω, the atomic volume. If the critical strain rate is less than the imposed strain rate then cavity nucleation is likely. It can be seen from the relationship that small grain sizes and small particles will minimise cavitation. It is also predicted that deformation at high temperatures (high D_{gb}) and/or slow strain rates will inhibit cavitation, but this is not commercially attractive since energy costs will be high and/or forming times will be too long. Additionally, structural instability at elevated temperatures, or over long times, can result in substantial grain growth and/or enhanced cavitation. Observations of cavities at the nucleation stage are difficult to make, and information on nucleation usually has to be deduced from specimens in which cavities have grown to a size where they are resolvable.

2.1 Hard Particles

Experimental studies on superplastic alloys have shown a correlation between the presence of hard second phase particles and cavitation, while microduplex materials which are particle free do not normally cavitate readily. Such a correlation has been observed for the case of the non-cavitating Pb-Sn eutectic [20]. The addition of a third element led to the formation of intermetallic phases of increasing hardness e.g. SbSn, Ag_3Sn and Cu_6Sn_5. On subsequent superplastic deformation, cavities were observed to nucleate at particle/matrix interfaces and this was attributed to the limited ability of the intermetallic phases to contribute to the accommodation of grain boundary sliding. The level of cavitation for a given strain was seen to increase as the hardness, volume fraction and particle size of the intermetallic phase increased. The presence of iron and silicon-rich inclusions in aluminium alloy 7475 [4], of primary $ZrAl_3$ and $CuAl_2$ in Al-Cu-Zr alloys (Supral) [43], of large particles of Ti(C,N) and σ-phase in α/γ duplex stainless steels [19,44], and of cobalt silicide particles, CoSi and $CoSi_2$, in the copper-base alloy Coronze 638, appear to be largely responsible for cavitation in these materials. Hard particles associated with cavities are seen in Fig. 4.

(a) (b)

Figure 4. Cavities associated with (a) $CuAl_2$ particles in Supral 220 (b) Ti(CN) in α/γ stainless steel IN744.

It is interesting to note that α/β titanium alloys do not readily cavititate. The impurities which are responsible for the formation of hard particles in other duplex materials such as carbides, nitrides and oxides in steels [19], and sulphides in α/β copper alloys [45,46], are not so important in titanium-based alloys which tend to take impurity elements into solid solution at elevated temperatures.

Particle/matrix interfaces on sliding grain boundaries are likely sites for cavity nucleation, since the stress concentrations and low surface energies combine to substantially reduce the critical void radius, r, equal to $2\gamma/\sigma$, where γ is surface energy and σ is stress [47]. In the absence of particles it is unlikely that the stresses and required vacancy supersaturations are attainable during deformation at elevated temperatures, and conventional nucleation by vacancy concentration is impossible [48]. However, since cavities cannot be observed at the nucleation stage, the association of cavities with intermetallic particles may also be the result of pre-existing defects such as cracks or regions of cohesion which have developed during the thermomechanical processing required to produce the superplastic microstructure [30].

2.2 Phase Proportions and Characteristics

In duplex materials the relative proportions and characteristics of the phases which make up the microstructure can have a marked influence on cavitation behaviour. This can be readily illustrated by reference to α/β brasses for which the volume fraction of cavities which form for given deformation conditions, increases as the volume fraction of α-phase is increased [49-52] Fig. 5. It has been proposed that the fcc α-phase behaves on a hard undeformable constituent and that grain boundary sliding is wholly accommodated by diffusion and plastic flow within the bcc β-phase. A reduction in the volume fraction of β-phase limits the ability of the microstructure to dissipate stress concentrations and hence promotes cavity nucleation. An analogous effect is observed on reducing the volume fraction of β-phase in the structurally similar α/β titanium alloys [25], although the levels of cavitation observed are appreciably less than for α/β copper alloys. For a number of α/β titanium alloys optimum superplastic behaviour is associated with a volume fraction of β-phase of about 0.4 [53]. This is presumably sufficient to accommodate grain boundary sliding and to give maximum stability to the fine grain microstructure.

2.3 Grain Size

There is considerable experimental evidence which supports the relationship between increasing grain size and increase in cavity nucleation. This observation may be interpreted in terms of the increasing distances over which the accommodation processes have to operate, or can be related to the associated increase in flow stress which will lead to a decrease in critical nucleus size. In the aluminium alloy Supral 220, where the volumetric growth rate is almost independent of strain rate and temperature, the volume fraction of voids was found to increase with increase in grain size [8] (Fig. 6). Moreover, grain growth during superplastic flow, has been invoked as a reason for the apparent continuous nucleation of cavities in both 7475 and Supral 220 [8,54]. The role of increasing grain size has also been reported for α/β brass [12] and for the highly superplastic Zn-Al eutectoid alloy [29]. For the latter material, cavitation was minimal for initial grain sizes <5μm. Superplastic deformation led to grain growth and when the grain size exceeded a threshold of about 8μm, nucleation of cavities became significant.

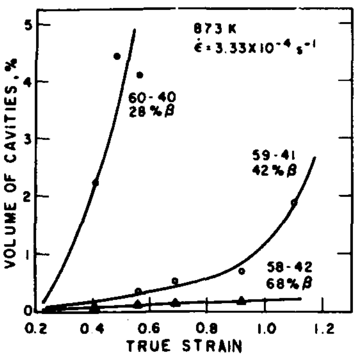


Figure 5. Cavitation in α/β brass as a function of superplastic strain for different phase proportions.

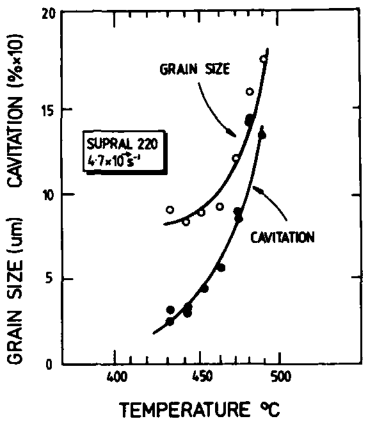


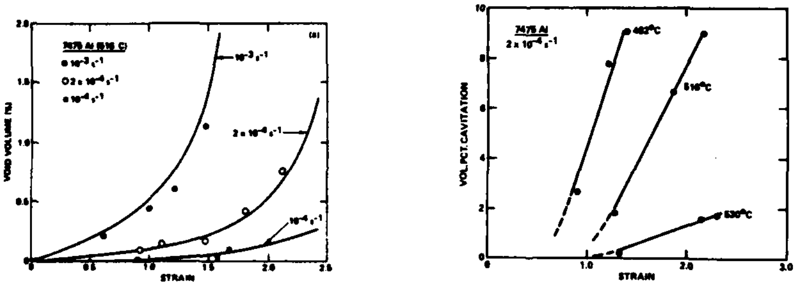
Figure 6. Relationship between grain size and cavitation for Supral 220 deformed to 300% elongation at various temperatures.

The importance of microstructural stability in relation to cavitation during superplastic flow has been reviewed for Fe-C alloys [16]. Plain carbon steels with a fine grain ferrite/cementite microstructure show extensive cavitation at Fe_3C/α interfaces and low elongations to failure, ~100% [55]. This is associated with coarsening of the cementite particles during superplastic deformation, which is accompanied by an increase in ferrite grain size. By increasing the carbon level of commercially based steels to ~1.6%, and hence markedly increasing the volume fraction of cementite, Sherby and his coworkers showed that a very fine stable microstructure could be produced on thermomechanical processing. The structure could be further stabilised by the addition of 1 to 1.5%Cr which partitioned preferentially to the cementite and increased its resistance to coarsening. The ultra high carbon alloys are capable of undergoing large tensile strains, >1000% elongation, and are extremely resistant to cavitation despite the increased volume fraction of cementite particles which they contain [56,57].

An analogous microstructure is that produced in a complex commercial aluminium bronze of nominal *composition:- Cu-10%Al-5%Fe-5%Ni*, by Higashi and coworkers [58]. The material can be thermomechanically processed to produce a fine grain microstructure (d ~2-3μm) stabilised by a dispersion of aluminides based on Fe_3Al and NiAl. At 800°C the alloy has an α-phase matrix containing about 30% by volume of β-phase and about 30% by volume of fine aluminide particles (K-phases). Its superplastic behaviour is untypical of that of most copper-base alloys in that it is capable of undergoing very considerable tensile strains with relatively low cavitation. The largest uniaxial tensile elongation so far recorded in the literature, 5,500% without failure, has been reported for this alloy [58].

2.4 Strain Rate and Temperature

It would be reasonable to expect that increasing the deformation temperature and/or reducing the strain rate would reduce cavity nucleation. Flow stresses would be generally lower and there would be more time at the lower strain rates for the relaxation of stresses generated by grain boundary sliding, while diffusion processes would occur more rapidly at the higher temperatures. However, the effect of changes in strain rate and temperature on cavitation during superplastic deformation varies widely depending on the material. The differences observed are often due to microstructural changes, particularly in the grain size of the material.



(a) (b)

Figure 7. Effect of (a) Strain rate and (b) temperature on cavitation in 7475 Al alloy [59].

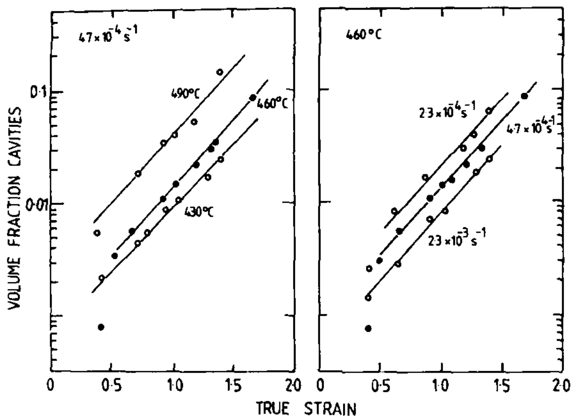


Figure 8. *Effect of temperature and strain rate on cavitation in Supral 220.*

In aluminium alloy 7475 the expected behaviour is observed in that increasing strain rate and decreasing temperature increase the level of cavitation for a given strain. This is illustrated in Fig. 7. for two batches of 7475 Al alloys showing different cavitation behaviours [59]. In Supral 220 decreasing the strain rate from $10^{-2}s^{-1}$ to $10^{-3}s^{-1}$ results in a decrease in the overall level of cavitation at a given strain but further reductions in strain rate to $10^{-5}s^{-1}$ result in a steady increase in cavitation [8] (Fig. 8). Since the strain rate sensitivity and volumetric strain rates are essentialy constant at strain rates from $10^{-5}s^{-1}$ to $10^{-2}s^{-1}$, it is unlikely that the increased levels of cavitation at the lower strain rates result from enhanced cavity growth. At the lower strain rates, grain growth during superplastic deformation is appreciable and it is likely that the detrimental effect of increased grain size on cavitation more than outweighs the benefical effect of lower strain rates. For Supral 220 an increase in deformation temperature leads to increasing cavitation through its effect on grain size [8] (Fig. 8 and Fig. 6).

Studies of cavitation in α/β *nickel silvers* have shown that both the volume fractions and size distributions of cavities are independent of strain rate and temperature [45] (Fig. 9). Similar observations on volume fractions of cavities have been reported for α/γ stainless steels [60] and for the single-phase copper-base alloy, Coronze 638 [15]. Since void growth during superplastic flow is strain dominated, as will be seen in the next section, the observed independence of the level of cavitation on both strain rate and temperature reflects the important role of pre-existing defects and/or the ease with which void nucleation may occur at second phase particles.

3. CAVITY GROWTH

3.1 Diffusion Controlled Growth

A cavity located on a grain boundary, whether nucleated during superplastic flow or pre-existing, may grow by stress directed vacancy diffusion along the boundaries which intersect the surface of the void, by plastic deformation of the surrounding material, or by a combination of both of these mechanisms, (coupled growth). Diffusion controlled growth of spherical intergranular cavities which are equi-sized, equi-spaced and small relative to the grain size has been analysed by several workers. The analysis of Raj and Ashby [47] gives

$$\frac{dr}{d\varepsilon} = \frac{D_{gb}\delta\cdot\Omega}{2kT\dot{\varepsilon}r^2}\left[\sigma_1{}^L - \frac{2\gamma}{r}\right]\frac{1}{\ln\left[\frac{\lambda}{2r}\right] - \frac{3}{4}} \qquad (2)$$

where $\sigma_1{}^L$ is the maximum principal stress local to the grain boundary, Ω is the atomic volume, γ the surface energy, r, the cavity radius and λ, the cavity spacing. Other relationships derived in the literature differ in the form of the last term, containing only r and λ. It can be seen from Eq. 2 that the rate of growth of an isolated void varies as the inverse square of void radius and thus slows down as the void grows.

Equation (2) has been derived for the case of uni-axial tension but can be written in more general terms which allow the effects of multi-axial states of stress and alternative geometries of deformation to be considered. The maximum principal stress, $\sigma_1{}^L$, can be re-defined in terms of the von Mises equivalent stress, σ_E, and the superimposed hydrostatic pressure, P. If the surface tension term in Eqn. 2 is ignored, since for the most part it is negligible with respect to σ_E, then

$$\sigma_1{}^L = \sigma_E\left[\frac{k_D}{3} - \frac{P}{\sigma_E}\right] \qquad (3)$$

where k_D is a geometric constant that depends on the mode of deformation (uniaxial, equi biaxial or plane strain) and the extent of grain boundary sliding.

Once the cavity becomes comparable in size to the grain size, d, of the material, either by coalescence with adjoining cavities or by the absorption of vacancies, the surface of the void will be intersected by several grain boundaries. It has been predicted that the rate of void growth due to diffusion along these boundaries then becomes:

$$\frac{d.}{d\varepsilon} = \frac{45}{d^2} \frac{D_{gb}\delta\Omega}{kT} \left[\frac{\sigma_1}{\dot{\varepsilon}} \right] \tag{4}$$

which unlike equation (2) is independent of cavity radius [61]

3.2 Plasticity or Strain Controlled Growth

Cavity growth as a result of deformation of the surrounding matrix leads to the relationship [62]

$$\frac{dr}{d\varepsilon} = \frac{n}{3} \left[r - \frac{3\gamma}{2\sigma_E} \right] \tag{5}$$

where n, the cavity growth rate parameter, is dependent on both the applied stress and the geometry of deformation. Unlike diffusion controlled growth, the rate of growth of voids increases linearly with void size and is independent of strain rate. Plasticity dominated void growth is governed by the ratio of the mean stress, σ_m^L, to the von Mises equivalent stress, σ_E. The form of n in equation (5) has been determined for rate sensitive materials to be [5,63,64]

$$n = \left[\frac{m+1}{m} \right] \sinh \left[2 \left(\frac{2-m}{2+m} \right) \left(\frac{k_s}{3} - \frac{P}{\sigma_E} \right) \right] \tag{6}$$

or [2,65]

$$n = \frac{3}{2} (1 + 0.932m - 0.432m^2)^{1/m} \left(\frac{k_s}{3} - \frac{P}{\sigma_E} \right) \tag{7}$$

where

$$\left(\frac{k_s}{3} - \frac{P}{\sigma_E} \right) = \frac{\sigma_m^L}{\sigma_E} \tag{8}$$

P is the superimposed pressure and k_s is a constant, equivalent to k_D, which is dependent on the geometry of deformation and the extent of grain boundary sliding. To obtain the values of k_s and k_D appropriate to uniaxial, balanced bi-axial and plane strain deformation, the stress state local to the sliding grain boundary needs to be determined.

From the analyses of Beere [40,66], and Cocks and Ashby [63,67], it can be shown that the extent of triaxiality on freely sliding grain boundaries is given by

$$\frac{\sigma_m^L}{\sigma_E} = \frac{1}{2} \left[\frac{\sigma_1^\infty}{\sigma_E} - \frac{\sigma_m^\infty}{\sigma_E} \right] \tag{9}$$

while for the case of *fully rigid* grains the local and remote mean stresses, σ_m^L and σ_m^∞, respectively, are equivalent. The magnitude of the maximum principal stress local to the grain boundary, σ_1^L, will also depend on the extent of grain boundary sliding. For the case of totally rigid grains the local maximum principal stress, σ_1^L, is equal to the remote maximum principal stress, σ_1^∞, while for a freely sliding boundary the ratio of the local maximum principal stress to the von Mises equivalent stress is given by

$$\frac{\sigma_1^L}{\sigma_E} = \frac{1}{2} \left[\frac{3\sigma_1^\infty}{\sigma_E} - \frac{\sigma_m^\infty}{\sigma_E} \right] \tag{10}$$

The relationships for the stress state local to the grain boundary apply only when the material is continuous and break down when cavitation occurs. They can only act as a guide to the magnitude of the maximum principal stress and the mean stress that might be experienced by cavities growing during superplastic flow. The limiting values of k_D and k_s may be obtained from the values of σ_1^∞/σ_E and σ_m^∞/σ_E derived from the remotely applied principal stresses in uniaxial, equibiaxial and plane strain deformation, by using equations (3),(8),(9) and (10). The values are summarised in Table I.

Table I. Limiting values of k_D and k_s for diffusion controlled and strain controlled cavity growth (0% = no grain boundary sliding, 100% = freely sliding grains)

	k_D		k_s	
Deformation Mode	0%	100%	0%	100%
Uniaxial	3	4	1	2
Equibiaxial	3	3.5	2	2.5
Plane strain	$2\sqrt{3}$	$2.5\sqrt{3}$	$\sqrt{3}$	$1.5\sqrt{3}$

The values of k_D and k_s appropriate to superplastic deformation will depend on the extent of grain boundary sliding. For most superplastic alloys about 50% of the strain is due to grain boundary sliding so that the value of either k_D or k_s can be taken as mid-way between the limiting values listed in Table I. It can be seen from equations (3) and (8) that cavity growth by diffusional or plasticity processes will cease when

$$P \geq k_{D,s} \frac{\sigma_E}{3} \tag{11}$$

Hence, growth of cavities can be prevented during superplastic forming by the application of a suitable confining pressure or "back"pressure.

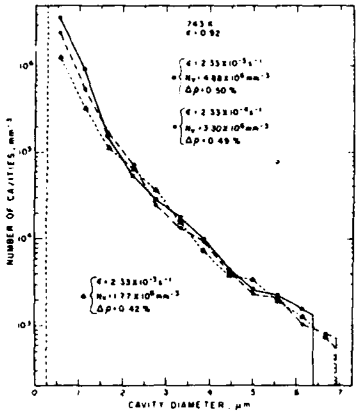


Figure 9. Cavity size distributions in an α/β Cu-Ni-Zn alloy deformed at three constant strain rates.

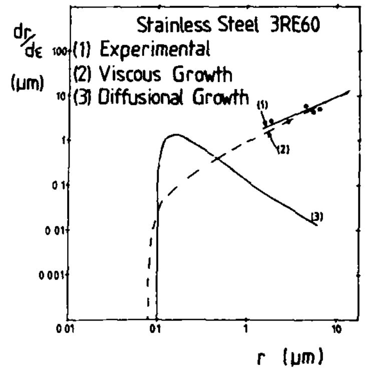


Figure 10. Comparison of experimental and predicted cavity growth rates for an α/γ stainless steel deformed in uniaxial tension at 1000°C.

3.3 Analysis of Cavity Growth

If the rate of change of void radius with strain predicted for each mechanism (equations 2, 4 and 5) is examined for the majority of superplastic materials deformed under optimum conditions, then the growth of individual voids would be expected to be dominated initially by stress directed vacancy diffusion (equation 2). However, when once the void attains a radius of ~1μm then strain controlled growth (viscous growth) dominates as is seen in Figure 10 for measurements made on an α/γ stainless steel deformed at 1000°C. Metallographic studies show that small voids (r < 0.5μm) usually have a spherical morphology, although it is uncertain whether this is the result of diffusional growth, or plasticity growth with surface diffusion, whereas larger voids may show more irregular morphologies (Fig. 11). Studies made of cavity closure rates during superplastic compressive flow show that shrinkage is also dominated by plastic flow [32]. Superplastic diffusional growth (equation 4) is only significant in materials with a fine grain size deformed at strain rates of about $10^{-4}s^{-1}$ or less, and then only for voids with diameters greater than the grain size and less than ~10μm.

The relationships which describe void growth (equations 2,4 and 5) each predict a different variation of void volume with strain. Void volumes in superplastically deformed specimens can be obtained using precision density measurements or quantitative metallography, and if plotted against strain should enable the dominant void growth mechanism to be identified, provided that cavity coalescence is not significant. For conventional diffusional growth, Eq. 2 predicts that void volume fraction should increase linearly with strain ($dr/d\epsilon \propto 1/r^2$; $V \propto \epsilon$). If the voids are larger than the grain size and diffusional growth dominates then the cube root of the void volume fraction will increase linearly with strain ($dr/d\epsilon$ = constant; $V^{1/3} \propto \epsilon$), whereas for strain controlled void growth the void volume fraction would be expected to increase exponentially with strain ($dr/d\epsilon \propto r$; $V \propto \exp(\epsilon)$).

The variation of the volume fraction of voids with strain has been measured for several aluminium alloys [5,34,68-71], α/β copper alloys [13,45,50,72,73], low alloy steels [18] and α/γ microduplex stainless steels [19]. In all cases the variation of void volume fraction with strain is best represented by an exponential relationship (Fig. 12). This supports the view that for the most part void growth is plasticity controlled. The most significant deviations from strain control occur at low volume fractions and low strains, where the voids are small and isolated, and where diffusion control might be expected to be important. However, few measurements of the variation of void volume fraction with strain have been reported for low strains.

4. CAVITY COALESCENCE

The change of void radius with strain calculated from equations 2,4 and 5 is plotted in Fig. 13a, for Supral 220 deformed under the optimum condition for superplasticity. It has been assumed that the void pre-existed and that at the start of deformation it had the minimum size consistent with thermodynamic stability ($r_o = 2\gamma/\sigma$). It can be seen that diffusion control is important in the early stages of growth. At a radius of ~0.6μm strain controlled growth is becoming important and dominates for void radii >1μm. Fig. 13b shows the calculated contribution of diffusion and plastic strain to the predicted (composite) growth rate. It can be seen that after strains of 2.0 (equal to 600% elongation) the largest void would only have a radius of 2μm, which is significantly less than that observed experimentally (Fig. 13b). These differences are attributed to coalescence which can make an important contribution to void growth in superplastic materials.

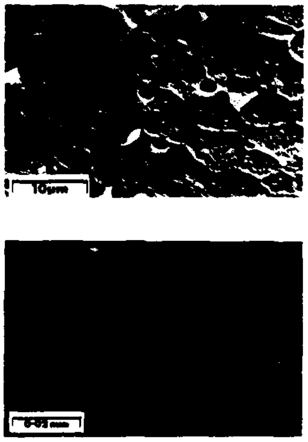


Figure 11. Cavity morphologies (a) Small spherical voids in Supral 220, ε = 0.41 (b) Large voids of irregular shape in 7475 Al alloy, ε = 1.06.

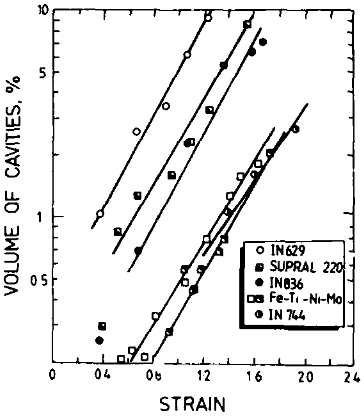
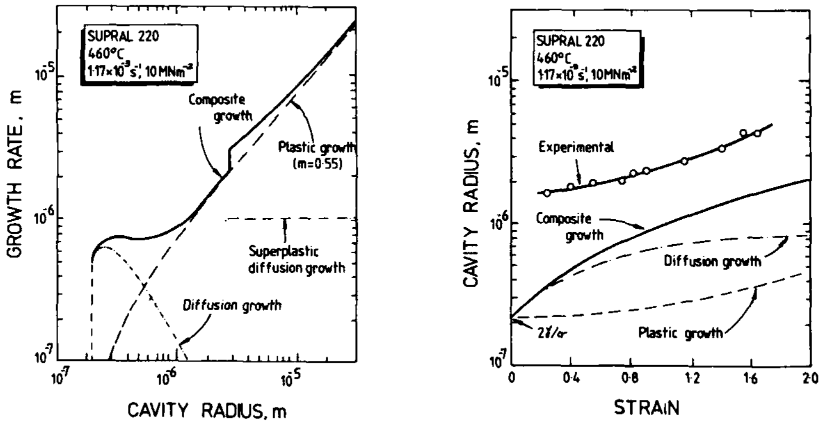


Figure 12. Variation of the volume fraction of voids (plotted logarithmically) with strain for several superplastic alloys.



(a) (b)

Figure 13. Supral 220 deformed in uniaxial tension (a) Calculated variation of void growth rate per unit strain versus cavity radius (b) Calculated and experimental variation of void radius with strain.

Metallographic evidence for the coalescence of cavities is available. The large elongated cavity seen in Fig.14 lying parallel to the rolling direction and the tensile axis is almost certainly due to the coalescence of cavities nucleated on closely spaced inclusions. Void coalescence therefore has the effect of extending the size distribution of cavities to larger radii and produces the large voids which are most likely to have a serious effect on post-forming properties. It is probable that some of the large voids which develop in aluminium alloys are associated with hydrogen outgassing. This leads to fine localised porosity which provides pre-existing sites for cavity growth and subsequent coalescence [69].

The extent of void coalescence du. . superplastic deformation depends on the size, population density and spatial distribution of voids wit.. . the material. Stowell et al [64] have examined the effect of pair-wise coalescence on the evolution of cavity size distributions in α/β nickel silvers. Reasonable agreement was observed between the size distribution measured at a higher strain and that predicted from the size distributi-n measured at lower strains except for the largest voids in the distribution. The discrepancies were attributed to a non-random spatial distribution of v ds, a feature which is frequently seen in cavitated specimens. Pilling [71] has examined the effect of pair-wise coalescence on the mean cavity growth rates for a 7475 Al alloy and has obtained good agreement between predicted growth rates and the measured values. In addition to confirming the importance of coalescence on the rate of accumulation of cavity damage in a single phase Cu-base alloy, Wilkinson and Caceres [74] have proposed that strain hardening due to grain growth can play a significant role in cavity growth and hence in coalescence.

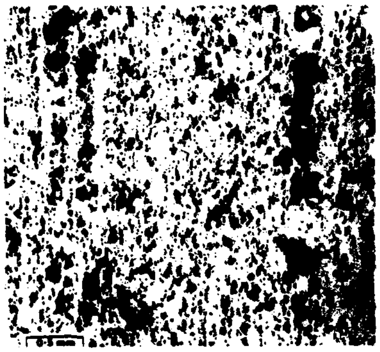


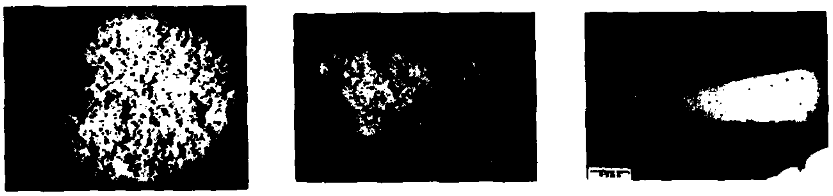
Figure 14. A heavily cavitated α/β Cu-Ni-Zn alloy showing substantial cavity interlinkage.

5. FAILURE OF CAVITATING SUPERPLASTIC MATERIALS

The growth, coalescence and interlinkage of cavities during superplastic flow leads to premature failure. It has been proposed that material between cavities acts as an internal neck and develops in a similar fashion to an external neck [75]. Final fracture in uniaxial tension occurs when successive separation of internal necks allow a transverse crack to propagate, resulting in a fracture face of significant cross-sectional area. Macroscopically the fracture face has a pseudo-brittle appearance, although microscopically it may show evidence of ductile tearing. However, in some cavitating superplastic Al alloys the final fracture face is microscopically intergranular. The criteria which determine the stage at which interlinkage occurs are unclear. Experimental data for α/β Cu-Ni-Zn alloys suggests that rapid interlinkage occurs when the cavity spacing is about twice the cavity diameter [72]. Stowell has proposed that cavities would join soon after their distance apart became less than the grain size [76,77], while for a single phase Cu alloy it has been reported that final fracture occurred when the reduction in load bearing section exceeded ~30% [15].

In non-cavitating systems, tensile specimens usually pull down to a fine point at fracture and fail by intrinsic plastic rupture at a localised neck. A gradual progression between the two extreme modes of failure can be observed for α/β brasses of varying phase proportions [50,52] (Fig. 15). In these alloys increasing levels of cavitation lead to fracture faces of increasing cross-sectional area. In the brasses, an alloy of high β content (42%Zn) is virtually non-cavitating and pulls down to a fine point at fracture. In contrast the 40%Zn alloy which contains a high volume fraction of α-phase fails without external necking after a relatively small elongation, <100%. The alloy of intermediate composition fractures by a mixture of plastic rupture and cavitation failure.

Lian and Suery [78] have recently developed an analysis for the quantitative prediction of the effects of both strain rate sensitivity and cavity growth on the fracture strain of superplastic materials deformed in uni-axial tension. The predicted strains for several alloys showed good agreement with those measured. However, the analysis does not take account of continuous cavity nucleation during deformation, cavity coalescence or grain growth, all of which can influence the strain to failure.



(a) (b) (c)

Figure 15. Longitudinal sections of α/β brasses pulled to failure (a) 60/40 (b) 59/41 (c) 58/42.

6. INFLUENCE OF HYDROSTATIC PRESSURE ON CAVITATION

Experimental studies have shown that annealing of cavitated material can reduce the volume fraction of voids [46,79]. However, the process is only effective in sintering small voids and the larger ones which have the most deleterious effect on post-forming properties are little affected. Post forming hot isostatic pressing also has the potential to remove completely voids formed during superplastic deformation [35,79,80], but is limited in application in view of its cost, its restriction on component size and the possibility that the voids will reappear on subsequent heat treatment [35].

From the previous section dealing with cavity growth it is clear that the superimposition of a hydrostatic pressure during SPF should restrict the rate of cavity growth. A simple criterion for the appropriate level of confining pressure was given in Eq. 11. The evidence summarised in Section 3 shows that strain rather than diffusion is the dominant void growth mechanism during superplastic flow so that K_s rather than k_D should be used in equation 11, although there are uncertainties in the appropriate values to use.

Quantitative studies of the influence of hydrostatic pressure on the accumulation of cavitation damage during superplastic flow are limited. Much of the available data relates to commercial Al alloys including Supral 220, 7475 and 8090. Bampton and co-workers [34,69] have reported on the effects of hydrostatic pressure on cavitation in 7475 Al alloy deformed under uniaxial, balanced biaxial and plane strain conditions. Subsequent studies have been made on 7475, 8090 and Supral 220 Al alloys, for both uniaxial and balanced biaxial deformation [5,70] (Fig. 16). In general increasing the superimposed pressure was found to:

(a) decrease the rate at which the volume fraction of voids increased with superplastic strain,
(b) decrease the level of cavitation for a given strain,
(c) displace to higher strains, the strain at which cavitation could be first detected,
(d) increase to a limiting value the strain to failure.

Recent work on aluminium alloys showed that the cavity growth rate parameter, n, was strongly dependent on the applied state of stress, and that cavitation was essentially eliminated when the imposed pressure was approximately equal to 0.5 of the von Mises equivalent stress (equal to the uniaxial flow stress) [5]. The value of 0.5 is similar to that of 0.6 reported by Bampton et al [69] and Story et al [70] for the elimination of cavitation in 7475 Al alloy deformed under optimum conditions. The variations of the void volume fractions with strain at different imposed pressures were consistent with strain controlled growth (Fig. 16). i.e. the volume of voids increases exponentially with strain. However, a variation in the measured volume fractions of voids was observed from specimen to specimen which introduced a degree of uncertainty into the measured values of n. This error, after normalising of the data, was too large to enable the uniaxial and balanced biaxial results for 7475, 8090 and Supral 220 Al alloys to be distinguished and equations (6) or (7) to be validated (Fig. 17). Further uncertainties in the measured values of n arise because the parameter relates to the growth of individual voids, where as the variation of the overall volume of voids with strain is influenced both by continuous nucleation and by coalescence. However, the values of n obtained from plots of log (void volume fraction) versus strain were greater for deformation in balanced biaxial tension that for deformation in uniaxial tension, an observation consistent with both equations (6) and (7).

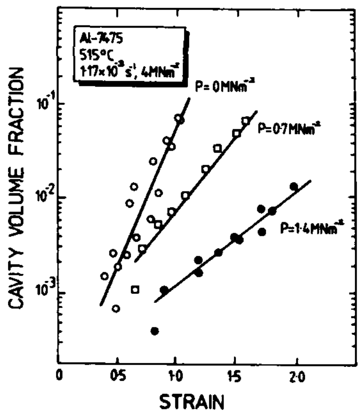


Figure 16. Measured variation of the volume fraction of voids with strain in 7475 Al alloy deformed in balanced biaxial tension with various superimposed pressures.

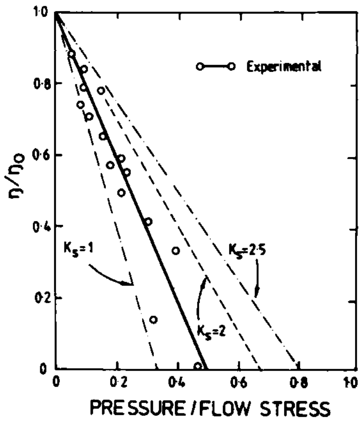
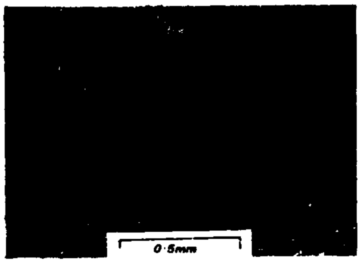
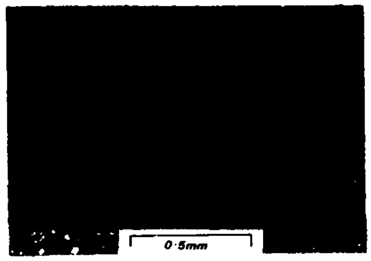


Figure 17. The experimental and theoretical variation of the cavity growth rate parameter, n, with pressure, plotted in normal-form.

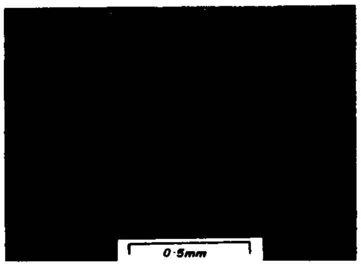
Quantitative optical metallographic studies of void size distributions which developed during deformation under various levels of imposed pressure showed that while the size to which voids could grow was markedly effected, (Fig. 18), there was little effect on the corresponding number of voids per unit volume (Fig. 19). The effect of the imposed pressure on void growth was accompanied by a change in void morphology. At zero and low imposed pressures voids which were initially spherical spread along the grain boundaries becoming trianguloid, as may be seen in Fig. 11b for Al alloy 7475. The spreading of the voids parallel to the grain boundaries permits coalescence to occur readily so that large voids can develop rapidly. At higher imposed pressures, the voids that were present retained their almost spher l form and as growth was limited, were unable to coalesce. Hence, the void networks observed at lower confining pressures were prevented from developing.



(a)



(b)



(c)

Figure 18. Cavitation in Supral 220 after uniaxial deformation to 450% elongation with superimposed pressures of (a) 0 MPa (b) 2.1 Mpa (c) 4.75 MPa.

Although the majority of commercially useful superplastic alloys have low flow stresses (<10Nmm^{-2}, ~1400 psi), it would be both costly and difficult to apply pressures greater than the flow stress to ensure that cavitation was totally eliminated. In practice, imposed, or back, pressures cannot normally exceed 3.5MPa (500psi) so the use of hydrostatic pressure is limited to slow strain-rate forming of materials such as 7475 and 8090 Al alloys where flow stresses at strain rates of 2 x $10^{-4}s^{-1}$ to 5 x $10^{-4}s^{-1}$ are of the order of 3 to 7Nmm^{-2}. However, the use of lower back pressures can limit void volume fractions to ~0.1%, and the voids which do form are small, spherical and relatively isolated. The effectiveness of back pressure in reducing the level of cavitation in an 8090 Al alloy is illustrated in Table II. The elongation to failure was increased from ~300% to > 600%, while the level of cavitation at 300% elongation was reduced from 5% by volume at zero imposed pressure to nil with an imposed pressure of 3.05MPa (450 psi).

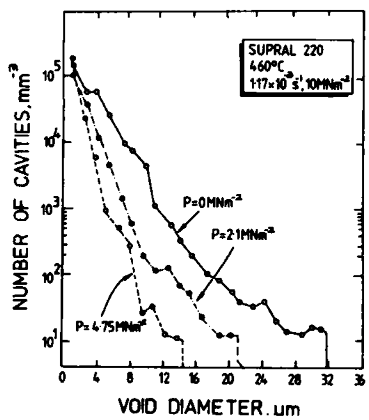


Figure 19. The influence of hydrostatic pressure on void size distributions in Supral 220 superplastically deformed in uniaxial tension to 450% elongation.

Table II. Effect of hydrostatic pressure on volume % cavitation in 8090 Al alloy deformed at 520°C and $10^{-3}s^{-1}$.

Strain % elongation	Pressure MPa 0	0.7	1.4	2.1	3.05
200	1.6	0.25	0.2	0.08	n/a
300	5.0	0.5	0.35	0.1	n/a
400	failed	1.1	0.55	0.11	0.02
600	failed	failed	failed	failed	0.12

7. CONCLUSIONS

There is currently a considerable volume of literature relating to both experimental and, to a lesser extent, theoretical aspects of cavitation during superplastic flow. It is clear that cavities are often associated with pre-existing defects, and with grain boundary particles. The ease with which a cavity will nucleate on a grain boundary particle will depend on its shape, size and hardness (reflecting the type of bonding) and on the ability of the phases adjoining the particle to contribute towards accommodation by diffusional or dislocation processes. Nucleation is less likely to occur the smaller the grain boundary particle size and the smaller the matrix grain, or phase, size. Cavity nucleation can be influenced to some extent by control of superplastic strain rate and temperature. However, in practice raising the temperature and lowering the strain rate to aid accommodation is not likely to be commercially attractive since both procedures may lead to grain growth and loss of superplasticity. In view of the instability and complexity of the microstructure of many superplastic materials, it is not surprising there have been no attempts to quantify the kinetics of void nucleation.

Experimental studies have shown clearly that void growth is essentially strain controlled. The development of large cavities at relatively small superplastic strains is likely to be the result of coalescence, or to the presence of relatively large pre-existing defects. In Al alloys large cavities may be associated with the outgassing of hydrogen. Further experimental and modelling studies on the role of coalescence on cavity growth are required. Although it is well established that the growth coalescence and interlinkage of cavities leads to a premature pseudo-brittle fracture, there have been very few experimental studies of the failure process. The development of a comprehensive analysis for the prediction of fracture strain looks formidable and would have to take into account the effects of strain rate sensitivity, cavity nucleation, cavity growth, coalescence, and strain hardening due to grain growth.

To minimise cavitation in superplastic alloys care must be taken at each stage in the production process with attention being paid to cleanliness of liquid metal, the absence of coarse primary phases during solidification, and the avoidance of fracture or decohesion of hard particles during thermomechanical processing to develop a fine uniform microstructure [30]. To prevent cavitation from occurring superplastic forming may be carried out with superimposed hydrostatic pressures of 0.5 to 0.75 times the uni-axial tensile flow stress. The condition for zero cavitation (zero cavity growth) depends solely on the geometry of deformation. For Al alloys such as 7475 and 8090 the pressures required to reduce cavitation to low levels at the strain rates and temperatures where optimum superplasticity is developed would be relatively low and manageable in many commercial forming operations.

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DIFFUSION BONDING OF METALS

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SUMMARY

The need to reduce the cost and weight of aerospace metallic structures has led to increased interest in solid state and liquid phase diffusion bonding processes, especially in combination with superplastic forming. The bonding mechanisms and bonding techniques are reviewed and the process variables that affect bond quality and strength are described with reference to bonds between Ti-alloys, Al-alloys and dissimilar metals. The importance of quality control and the limitations of current NDE techniques for diffusion bonding are emphasised. Finally some trends and priorities in diffusion bonding technology are indicated.

1 INTRODUCTION

A diffusion bonded joint in Ti-6Al-4V alloy is an engineer's dream - it can have the microstructure and mechanical properties of the base metal. Such joints are currently manufactured as part of the superplastic forming and diffusion bonding (SPF/DB) of titanium aerospace components[1-4]. Unfortunately high quality diffusion bonded joints in other alloy systems can be more difficult to achieve and require a detailed understanding of the mechanism of bonding and of the metallurgy of the alloy. Excellent general reviews of diffusion bonding (db) techniques and applications have been presented by Schwartz[5] and Kazakov[6] whilst Stephen and Swadling[3] have reviewed the application of db to aircraft structures.

In view of the confusion in the literature[5-7] a brief summary of the preferred terminology for diffusion bonding[8] is necessary. Diffusion bonding refers to a joining process that requires high temperatures to enhance diffusion but involves little macroscopic deformation; the joint formed without interlayers has the microstructure and composition of the base metal. Diffusion bonding can be divided into:

i diffusion welding, the joining of two metals by pressure and temperature without melting. An interlayer foil or coating may be used. This process is sometimes referred to as solid state db or hot press bonding.

ii diffusion brazing, the joining of two metals by heating to produce liquid at the bond interface via an interlayer foil or coating. Pressure may or may not be used. This process is sometimes referred to as transient liquid phase db, activated db, eutectic db or liquid phase db.

Some characteristics of diffusion bonded joints[4,5-7,9] are listed in Table 1. Since the process time does not depend on the complexity of the structure or on the bond area or on numbers of components, there is potential for reduced costs with increase in size and numbers of components per bonding cycle[3,4]. The advantages are increased when superplastic forming and diffusion bonding are combined to produce titanium sheet structures[1-4] and db may be the only practical method available for the manufacture of metal matrix composites[3].

Experience gained in bonding Ti-alloys has emphasised the need for a quality control and NDE capability for db joints and for a more quantitative approach to the effect of defects on the mechanical properties of the joints. These aspects are also becoming of increasing importance in the db of Al-alloys and metal matrix composites. This lecture therefore concentrates upon those features of the db process that can give rise to defects and affect the mechanical properties of the joint. Reference is made primarily to Ti and Al-alloys as these are typical of alloys that are easy and difficult to bond respectively. Particular attention is paid to bonding and testing techniques used on a laboratory scale since these form a vital part of any process development for db. Finally a brief mention is made of the limitations of current NDE methods as applied to db joints.

2 MODELS FOR DIFFUSION BONDING

Diffusion bonding in the solid state involves the following mechanisms:

i plastic deformation on loading) Stage 1
ii creep deformation)
iii diffusion) Stage 2
iv recrystallisation and grain boundary migration)

At low pressures (much less than the macroscopic yield stress) and high temperatures (>0.5 T_M where T_M is the absolute melting point) used in bonding the deformation is confined primarily to the surface asperities as shown in Fig 1[10]. Stage 1 involves instantaneous deformation on loading followed by power law creep according to a relationship of the type:

$$\dot{\epsilon} = A\sigma_e^n \exp(-Qc/RT) \qquad (1)$$

where $\dot{\epsilon}$ = creep rate, σ = stress, Qc = activation energy for diffusion and A and n are constants with n ∿ 3-4 for Ti-6Al-4V[11]. For fine grained material superplastic deformation with little strain hardening (n ∿ 1.5-2.0) can accelerate the deformation of the asperities in Stage 1[12,13]. In UHC steel a decrease in grain size from 2 to 0.4 μm was predicted to increase the strain rate by a factor of 1000[13].

At the end of Stage 1 the bond interface consists of bonded areas separated by areas containing small voids (Fig 1). Hydrostatic pressure can accelerate void closure by diffusion and plastic deformation for voids >20 μm [10,14], but below this size diffusion alone controls the elimination of voids by surface, grain boundary and volume diffusion mechanisms. Fast grain boundary diffusion is favoured by a small grain size but in practice only time and temperature are important variables for many metals[15]. Small empty voids may be removed by isothermal anneals in the absence of pressure[7,16]. Stage 1 bonding is sensitive to stress and occurs much faster than bonding in Stage 2. The predicted contributions of creep deformation (mechanism 2) and of diffusion (mechanism 3) to the bonding of copper is illustrated in Fig 2 from the work of Nishiguchi and Takahashi[17]; at low pressure (3 MPa) the diffusion mechanism dominates but at 8 MPa the creep deformation mechanism dominates at all temperatures above about 950°K. Mechanism 4 which operates in the final stages of bonding may be essential for high strength joints since it leads to the elimination of the planar grain boundary interface[18,19]. Diffusion bonding models based upon an idealised surface and the above mechanisms have been reasonably successful in predicting the effect of pressure on the time to produce a 95% parent metal bond strength in titanium and copper[12,20,21]. More complex models involving evaporation and condensation of atoms may be required to explain bonding in the presence of oxide films[18,22].

Interlayers in the form of metallic foils or coatings (electroplated, evaporated or sputtered) are widely used in db joints[5]. They can lead to reduced values for the bonding parameters (temperature, pressure, time) and involve diffusion welding or diffusion brazing. The interlayers must be carefully selected to avoid changes in microstructure or composition which adversely affect mechanical and corrosion properties.

When the temperature or pressure required to obtain good surface contact prohibits diffusion welding, diffusion brazing offers an alternative joining method. This process has some of the advantages of conventional brazing (low pressure, short times) and of diffusion welding (base metal microstructure and strength). Diffusion brazing has been described by Owczarski[23] using Fig 3. An interlayer or coating with composition 1 is heated to the bonding temperature and melted. During isothermal annealing diffusion changes the composition from 1 to 3 in Fig 3 and the joint solidifies. Further diffusion results in a large dilution of the interlayer elements and minimal effect on the parent metal. This technique has been applied to both titanium and aluminium alloys.

Interlayer coatings may also be removed by diffusion entirely in the solid state. A section through a bond made between 1 μm thick silver coated clad Al-alloy sheet is shown in Fig 4a; the Al-Ag phase diagram shows that silver will form a solid solution in aluminium during isothermal anneals at 500°C and given sufficient time at temperature the Ag layer can be completely removed (see para 8).

3 EFFECT OF SURFACE ROUGHNESS

For a given set of processing parameters, surface roughness is probably the most important variable affecting the quality of diffusion welded joints. The roughness affects the time required to achieve complete contact between the surfaces being bonded. In practice a metal surface has asperities which are small in terms of height and wavelength (surface roughness) superimposed on longer wavelength asperities (surface waviness) as shown schematically in Fig 5[24]. Surface roughness as normally measured refers to the short wavelength asperities and typical values for various surfaces are shown in Fig 6[3]. The long wavelength asperities are particularly important in the bonding of thick section machined parts ie massive db. In these joints regions of high porosity can be separated by completely bonded regions and the porosity present may be further apart and greater than expected from the surface roughness data[10]. Massive db imposes severe requirements on the accuracy and alignment of the tooling and may require loads to be applied in more than one direction[3]. Massive db therefore tends to be more difficult and costly than gas pressure bonding of thin sheet.

Thin sheet has the advantage that in the as rolled state the surface finish is usually good. Talysurf traces for a clad Al-alloy sheet in the as rolled state and after polishing on 1 μm diamond paste or grinding on 600 silicon carbide paper are shown in Fig 7[25]; silicon carbide particles were found embedded in the ground surface (Fig 4b) and the strength of diffusion bonded Al-alloy joints decreased with increase in roughness (Fig 8)[25,26]. The finish obtained for titanium alloy sheet in the as received state or in the carefully ground state is typically better than Ra = 0.5 μm. A ground surface of Ti-6Al-4V alloy is shown in Fig 9a; note the occasional deep grooves and surface debris from the grinding operation. The initial stages of bonding of this surface were studied by fracturing the bonded joint after very short bonding times of ∿4 min[27]. The fracture showed the bottom of the grooves remained unbonded (Fig 9b), the bonded asperities exhibited ductile fracture cusps (Fig 9c) and regions that had been deformed but not bonded appeared flat and without ductile cusps (A in Fig 9d). The latter regions are typical of intimate contact disbonds caused by contaminated interfaces or incorrect bonding conditions and are impossible to detect by current NDE techniques.

Surface roughness is responsible for the residual porosity at the end of Stage 1 bonding and for the variation in void size. Surface grooves can give rise to channels and large and small voids (Fig 10b) and local waviness can lead to elongated voids or wide channels (Fig 10a)[27]. In titanium alloys microstructure also affects the amount of porosity present in the bond interface. The void ratio (ratio of unbonded/bonded area) for db Ti-6Al-4V has been measured by Enjo et al[28]; the alloy was first annealed in the α-β phase region (900°C, 4 h) or in the β-phase region (1010°C, 1 h) to produce coarse grains (∿8 μm) or coarse α platelets respectively compared with the as received material (2-3 μm grain size). The as received and 900°C annealed material showed increased void content with increase in roughness, but for the 1010°C annealed material the void content remained high and insensitive to surface finish (Fig 11). These void contents reflect the difficulty in producing good surface contact with coarse or irregular microstructures and may explain the variability in bond quality reported in the literature[29].

It is not always appreciated that surface roughness of thin sheet is proportional to the product of initial grain size and plastic strain[30]. The increase in roughness ΔRa for commercial purity titanium deformed in the temperature range 20-450°C is shown in Fig 12; the roughness increases with temperature and with strain. Surface roughness is also increased during superplastic deformation[15] and was associated with grain boundary sliding, as shown in Fig 13 for a Ti-6Al-4V alloy test piece after 300% extension at 925°C[31] After strains of ∿1.3, the Ra values increased from 0.7 μm to 1.5 μm for Ti-6Al-4V and from 0.3 μm to 2.7 μm for Al-Li-Cu alloys. These increases in surface roughness may increase the adhesion of contaminants and the reactivity of the surface.

4 CONTAMINATION AND DEFECTS IN DIFFUSION BONDED JOINTS

The most common surface contaminants encountered are inorganic or organic films[32] and these may be reduced by solvent cleaning and pickling[15,33]. Organic films (oil or fingerprints) and water vapour may be removed by heating to ∿300°C but at higher temperatures surface reactions can occur[7]. These can be important for Al-alloys since stabl oxide films may be produced[34,35] which seriously reduce bond strength[26,36]. At high temperatures many metal oxides dissociate eg Ag_2O or dissolve in the base metal (oxides of Cu, Ti, Zr, Nb, Ta, Nb, and Ni). The rate of dissolution of 100 Å thick oxide films is virtually instantaneous for all these metals at 0.5 T_M except for Ni which requires ∿0.92 T_M; the time to dissolve an oxide film increases rapidly with increase in oxide film thickness[37]. Stable oxides, carbides or nitrides may be difficult to remove except by sputter cleaning at low temperatures[32]. Particular problems may arise in SPF/DB processing of titanium alloys[39]. The defects found may be divided into[3];

i Large voids
ii Microvoids
iii Intimate contact disbonds

Large voids or disbonds are likely to be associated with argon gas entrapment and can be avoided by progressive venting[3]. In massive db similar voids are produced by poor surface finish or tool misalignment. Microvoids (Fig 10) can indicate incorrect bonding parameters. Intimate contact disbonds can be caused by surface contamination or oxide films and are the most difficult to detect because of their limited thickness. These films can however lead to very low bond strengths.

Although thin oxide films on Ti-alloys are readily removed by dissolution at the bonding temperature (∿925°C) further oxygen pick-up from the environment can seriously effect the quality of the bonds. Increasing oxygen in solid solution significantly increases the hardness[40] and excessive pick-up produces a hard oxygen-rich stabilised α-phase surface layer (α-case) which prevents bonding[19] and leads to surface cracking. The argon gas used to flush or pressurise the die chamber can be contaminated by residual O_2, N_2 and H_2O or by outgassing of stop-off compounds. The stop-off compounds (BN, Y_2O_3) may also become less adherent at high temperatures as the binder is burnt off and they may then become displaced on to the surfaces to be bonded[15,41].

5 BONDING TECHNIQUES

Diffusion bonds are made either as part of an investigation to develop a d method or to manufacture a component by an established technique. In the first case the need is for a d' that can be tested whereas in the second case the primary requirement is for an efficient structure. A s ple low cost db butt joint produced with bar stock allows standard tensile and fatigue test pieces to be manufactured with the bond interface at the centre of the gauge length. In tensile tests on these joints the bond is loaded uniformly and poor bonds may be reflected in low reduction of area values. The bond quality can vary across the bonded area however, especially if the faces to be bonded are not normal to the axes of the cylinders or if axial alignement is poor. To limit the deformation across a bond a spacer can be used[42] (Fig 14). In the bonding of unconstrained cylinders the maximum bonding pressure may be limited by bulging, which may be quite different for each metal in dissimilar metal bonds as shown for a Ti-6Al-4V/stainless steel bond in Fig 15.

Metals in thin section can be bonded under gas or platen p re to produce overlap shear test pieces. The strength of overlap shear test pieces is sensit. overlap length and to out of plane stresses which cause bending of the test piece. A test piece used for bonding Al-alloys[9,43] is shown in Fig 16. It requires 2 blanks (25 × 35 mm) which are bonded in a jig (Fig 17) to ensure the blanks are parallel after bonding (Fig 18). Cut-outs at A and B in Fig 16 provide metallographic samples after bonding and after heat treatment. However a planar bonded interface may not be produced over the whole overlap area since local formation occurs at the ends of the bond at B in Fig 18b with unbonded regions at C. Too small an ove p produces an inclined bond plane with adjacent shear leading to recrystallisation of the base Al-allo Fig 19).

Diffusion bonds in components can be divided into those made by massive db and those made between thin sheet. Massive db involves the joining of thick section machined parts under relatively high pressure (14 MPa, 2000 psi[19]) applied by mechanical means. Japanese workers[44] have reported on a wide variety of Ti-6Al-4V alloy aircraft components diffusion welded in vacuum (10^{-3} Pa) under a lower pressure of 2.94 MPa at 900°C for ∿2 h. Massive diffusion bonded products tend to be substituted for complex machined or forged components to improve material utilisation or cost and may therefore be heavily loaded.

Thin Ti-6Al-4V alloy sheet can be diffusion bonded by gas pressures of ∿2 MPa (300 psi). The advantage over massive db is that the pressure acts normal to the sheet surface whatever the shape being bonded and large area bonds with intimate contact are easily obtained. A pack bonding process can be combined with gas pressure superplastic forming.

Platen pressure bonding followed by gas pressure free forming of hemispheres[9] is shown schematically in Fig 20 and can be compared with 2 sheet pack bonding and forming into shaped dies in Fig 21[3]. More complex 3 and 4 sheet SPF/DB structures are shown in Figs 22-23. Note that in Fig 21-22 all the bonds are made initially by pack bonding (primary bonds) whereas in Fig 23 some bonds are made later after SPF (secondary bonds). The surfaces for the secondary db may therefore be exposed to the pressurising gas environment longer than the primary bonds with a greater risk of surface contamination.

The need for greater stiffness and strength at elevat ' temperatures has led to the most complex db components to be proposed for turbine blades (Fig 24)[45]. These may involve thick and thin sections, dissimilar metals and metal/ceramic interfaces and hot isostatic pressure techniques may be used[46].

6 TESTING OF DIFFUSION BONDED JOINTS

The strengths reported for db joints show a wide variation depending on bonding parameters, bonding technique and test piece design. The latter is particularly important since a test piece must provide meaningful strength data to enable the effects of metallurgical and processing variables on bond strength to be assessed.

A round bar test piece provides the best test conditions for a diffusion bond, but for thin sheet unsupported overlap test pieces are often used. In the latter under tensile load the stresses are greatest at the ends of the bond region and joints tend to fail under peel stresses (Fig 25)[47,48]. To reduce bending the Al-alloy db test piece in Fig 16 was tested in tensile shear in a constraining jig (Fig 26)[43]. For gas pressure bonded thin sheet an overlap test piece can be produced by machining two flat bottomed grooves[49] as shown in Fig 27; the bottom of the grooves should coincide with the centre of the bond. Typical load-time curves for constrained test pieces are shown in Fig 28 a-b; good bonds exhibited some plasticity whereas poor bonds failed with little plasticity.

Joint strengths have been compared in terms of the joint efficiency (JE); JE = P x 100/σA, where P = maximum load, σ and A are the tensile strength and cross-sectional area of the sheet respectively[50]. Failure in the parent sheet gives JE = 100%. The parameter JE does not require bond shear fracture and is therefore not a good measure of bond quality. A more satisfactory measure of bond properties is obtained by measuring the bond shear ratio $R_B = \tau_B/\tau_p = P_B/P_p$ where τ_B and τ_p are the shear strengths of the bond and parent metal respectively and P_B and P_p are the corresponding failure loads under shear conditions. It was found that for diffusion bonds between 3.2 mm thick Al-alloy sheet and for increasing overlap L, JE increased and R_B decreased[48]. For L in the range 2-6 mm, τ_B was approximately linearly related to L and a difference in L of 1 mm changed τ_B by $\sim$ 13 MPa (8%) (see para 8). It is often difficult to control the overlap to within 1 mm and this contributes to the scatter in shear strength data.

The fracture appearance of db joints can provide information on the quality of the initial bond[33,51]. For example incorrect bonding parameters may lead to unbonded regions (Fig 9), but a good bond will reveal uniform ductile shear and tensile fracture cusps as shown in Fig 29.

A peel test measures the resistance to crack growth at the bond line. The test is difficult to perform in practice[49] and is associated with considerable scatter, but the peel strength may be a useful indicator of a poorly bonded joint. Peel strength values are useful only for comparing identical test pieces. A typical peel load - time curve is shown in Fig 28c; the load plateau corresponds to stable crack growth[9,18].

7 DIFFUSION BONDING TITANIUM ALLOYS

The bonding parameters for Ti-alloys have been reported by many workers[3,5,6,19]. The creep rate increases above about 850°C and decreases rapidly as the β-transus is approached and grain growth occurs (Fig 30)[52]. The creep rate depends on grain size and microstructure and these variables therefore effect diffusion bonding. The effect of grain size on the time and pressure required to produce a pore free bond in Ti-6Al-4V alloy is shown in Fig 31,[19]; at a typical bonding pressure for this alloy (300 psi, 2 MPa) the bonding time is increased by a factor 6 when the grain size increased from 6.4 μm to 20 μm. Since the grain size is coarser in plate or forgings than in sheet, different bonding parameters will be required for these different product forms.

The effect of prior anneals in the α + β (900°C) and β (1010°C) phase fields on creep rate and diffusion bonding of Ti-6Al-4V has been measured[28]. Creep rate at 850°C decreased with increase in annealing temperature due to grain coarsening or the presence of acicular microstructure (Fig 32); note the sensitivity to stress. Bonding was also much more difficult for the β-annealed material compared with negligible porosity for the as received material. In addition the bond strength increased more rapidly with increase in bonding temperature for the as received fine grained material. Heating to the β-transus temperature after intimate contact was obtained produced good bonds by causing grain boundary migration at the bond interface[19]. The effect of surface finish on bond porosity is discussed in para 3 and shown in Fig 11. An improvement in surface finish led to a dramatic increase in bond strength, but the 900°C annealed material remained inferior to the as received material (Fig 33). It was concluded that the void ratio and hence the bond strength was dependent on the deformation only (Fig 34) and that accelerated bonding could be obtained under superplastic conditions. Bonding Ti-alloys under superplastic conditions was reported[53] to reduce the bonding pressure by a factor 4, the welding time by a factor 6-30 and the temperature required by 50-150°C.

The effect of defects on the mechanical properties of diffusion bonded joints is of great practical importance, particularly as it is well known that bond defects which have little effect on bond tensile or shear strength can affect other mechanical properties such as fatigue or impact strength. Early work also showed that bonded joints with parent metal strength could be associated with ductility less than in the parent metal (Fig 35)[54].

The reduction in void content in the diffusion bonded joints of 2 titanium alloys with increase in time at 925°C and 0.52 MPa pressure is shown in Fig 36[27]. The appearance of the voids after ½ h is shown in Fig 10; parent metal tensile strengths were obtained for these bonded joints but there was a marked decrease in the fatigue limit (Fig 37a). Improved bonding conditions led to bonds with parent metal fatigue strength (Fig 37b), but the bond impact strength values were lower than for the parent metal[27,55] and were not significantly changed with increase in bonding pressure[55] (Fig 38). Impact properties appear to be more sensitive to bond quality than other mechanical properties[44].

There is increasing evidence that the bond strength in titanium alloys is dependent on the deformation history of the surface asperities. Russian workers have shown[56] that more rapid bonding can be obtained by cycling the bonding pressure. For example Ti-6Al-4V alloy at 850°C was cycled to 90% of the yield stress (σ_{ys}). After 10 cycles in 5 min the relative bond strength bond/parent $\bar{\sigma}$ = 0.8 and after 80 cycles in 20 min $\bar{\sigma}$ = 1 (curve 1 in Fig 39); the bond strength increased more slowly under a constant pressure of 0.9 σ_{ys}

(curve 2 in Fig 39) and the macroscopic plastic strain increased from 0.6% (pressure cycled) to 4.5%. These results were attributed to more rapid deformation of the asperities during pressure cycling.

Further work has shown that for bonding deformation below 5% the relative impact strength (strength of bond/strength of base metal) is independent of strain rate (Fig 40a) but at 5% deformation and above maxima in the relative impact strength curves were obtained in the superplastic strain rate range for titanium and nickel base alloys (Fig 40 b,c). However the tensile strength and ductility could be less than for the base metal[57,58]. It was concluded that good interface contact was obtained, but at the low superplastic strain rate strain hardening of the asperities was not sufficient to cause recrystallisation and a planar grain boundary interface resulted. At higher strain rates interface contact was poor due to strain hardening of the asperities, but where contact did occur recrystallisation and grain boundary migration produced a high strength bond interface. To achieve both high impact and tensile strengths it was suggested that strain rate or temperature should be changed during the bonding operation[59].

Interlayers used for bonding Ti-alloys include soft commercial purity titanium foils for diffusion welding and electroplated coatings for diffusion brazing[5]. The latter is a viable commercial process as demonstrated by the production of hollow titanium alloy fan blades using electroplated Cu and Ni interlayers[60]. Coarse grained Ti-alloys could be bonded under low pressure to give parent metal strength by using a fine grained superplastic Ti-alloy interlayer[59].

8 DIFFUSION BONDING Al-ALLOYS

Diffusion bonding of these alloys has proved difficult because of the tenacious surface oxide film[5,6]. Large deformations and high temperatures, in for example roll bonding[6,49] disrupt the oxide film but the introduction of soft interlayers in the form of cladding, coatings or foil inserts enable bonds to be made with smaller overall deformation. The surface coatings produced by ion-plating appear superior to those produced by electrodeposition, chemical vapour deposition or plasma spraying[33,61-63]. The development of superplastic Al-alloys and Al metal matrix composites has increased the need for a diffusion welding or diffusion brazing method for high strength Al-alloys[4,61,63,64].

Byun et al[33,61] have successfully bonded 7475 Al-alloy sheet using the 5052 Al-Mg alloy as an interlayer at 500°C and 2.76 MPa pressure; the effect of different interlayers on bond shear strength is shown in Fig 41. The bond shear strength increases with increase in deformation from 3% to 15% and with increase in bonding time up to 60 m (Fig 42). Bond shear strengths of 172 MPa and 241 MPa were reported for the as bonded and T6 heat treated conditions.

To reduce the bonding pressure, diffusion brazing has been tried with a variety of interlayers eg Ag, Cu, Zn,/Cu, Sn/Ag, Al-Si, Mg and Al-Si-Ge[61,62]. The additional interfaces can increase the oxide content and too thick an interlayer produces excessive concentrations of alloying elements which can lead to the formation of intermetallics and brittle bonds. However, for a total braze interlayer thickness of ∿ 25 μm, the bond shear strength approached that of the 7475 base metal[61].

Another possibility is the production of oxide free surfaces on Al-alloys by argon-ion sputter cleaning and coating with silver[25]. Such surfaces are readily bonded above the dissociation temperature of Ag_2O (∿200°C) and in this respect the bonding operation assembles that for titanium alloys. To avoid the formation of intermetallics and reduce the bonding pressure clad sheet can be used; a typical bond produced between 1 μm thick Ag coated and clad 7010-alloy sheet at 450°C and 7 MPa pressure is shown in Fig 4a and 43a. After solution heat treatment for 16 h at 480°C the silver layer was replaced by small grains (Fig 43b) and the silver concentration at the bond interface was reduced to about 1% (Fig 44)[9]. The bond shear and peel strengths were dependent on alloy composition and heat treatment (Fig 45)[65]. The clad layer allows extensive plastic deformation to occur in the SHT condition as shown in the partly cracked peel test piece in Fig 46; much less plasticity was apparent after ageing. As for Ti-alloys a measure of the bond quality and ductility can be obtained from bond fractures[48,61]. For the clad 7010 Al-alloy in the SHT condition ductile shear fracture was obtained (Fig 47a) and after ageing (Fig 47b) the fracture was much smoother.

Minimum interlayer thickness is required to ensure composition changes are not detrimental. The combination of high solubility and high diffusion rates enables Mg and Zn to diffuse rapidly into Al-alloy interlayers but the composition and strengthening effects depend on interlayer thickness and diffusion times. For example even after long solution heat treatment times (16 h) which increased the solute content and solid solution hardened a 2 x 120 μm thick clad interlayer between 7010 Al-alloy (Fig 48), a composition trough persisted (Fig 44); this prevented age hardening of the clad layer (Fig 48)[9]. However for a 25 μm thick 5052 interlayer between 7475 Al-alloy[33] the zinc composition trough was eliminated during the 60 min bonding operation (Fig 49).

Thus the concentration of elements that diffuse more slowly eg Ag or Cu, may be insufficient to produce age-hardening in thick interlayers and a concentration peak may remain at the interface after long diffusion times (Fig 44). Excessive alloy concentrations introduced by interlayers can lead to intermetallic formation. For example increasing the Ag concentration by decreasing the clad layer thickness for a fixed 1 μm thick Ag coating causes a reduction in bond strength (Fig 50) and at a clad layer thickness of ∿ 30 μm a continuous intermetallic layer was obtained [25] (Fig 51).

Although the above bonding techniques offer the possibility of platen or gas pressure bonding of Al-alloys, the elevated temperature bond peel strength may be too low for superplastic forming of 3 or 4 sheet structures[48].

The testing of db joints between Al-alloys can present problems[48]. Because of the ductility of Al-alloys, bending of overlap shear test pieces (Fig 25) causes local, high peel stresses which produce an apparent decrease in the shear stress values with increasing overlap length. Test pieces sheared in a restraining jig produced higher strengths (Fig 52a) than unrestrained test pieces (Fig 52b) and aged material was particularly sensitive to peel stresses (Fig 52b). It is therefore difficult to compare the strengths of Al-alloy diffusion bonded joints obtained under different overlap or test conditions.

9 DIFFUSION BONDING DISSIMILAR METALS

The bonding of dissimilar metals is likely to increase in the future with the trend towards aerospace structures made of different metallic and non-metallic materials. All the previous comments on bonding, especially on the use of interlayers are applicable to dissimilar metal bonds, but greater restrictions are imposed by the need for physical and chemical compatibility[36]. The use of multiple interlayers is more common for dissimilar metal bonds and ideally the interlayers should exhibit mutual solubility without intermetallic formation, a wide temperature range for bonding (0.53-0.7 T_M) and compatible thermal expansion and Young's modulii. Thin interlayers enable solid solution hardening to occur and optimum strength has been related to the interfacial strengthening.

High strength solid state diffusion bonds between Ti-alloy and stainless steel[55] provide a good example of the type of interface needed for dissimilar metals. Bonds between the base metals would form the intermetallics TiFe, $TiFe_2$ or $TiCr_2$ unless a low bonding temperature, short times and high bonding pressure was used, which is impractical. One solution is therefore to make a bond composed of stainless steel/nickel/copper/vanadium/Ti-6Al-4V interfaces. A bond made at 850°C under 10 MPa pressure for 1 h is shown in Fig 53; a tensile strength of ∿ 400 MPa and 460 MPa was obtained for Ni/Cu/V and Cu/V interlayers respectively, approximately 67% and 77% of the tensile strength of the stainless steel, but impact strength was low.

Bonds between dissimilar metals, metal matrix composites or rapidly solidified powder products may also be required in the future[66] and the processing temperatures and total deformation may need to be tightly controlled to preserve microstructures and properties. Diffusion welded or brazed joints in metal matrix composites have shown considerable scatter and low joint efficiency attributable both to processing and testing techniques[50,64]. An example of a proposed[50] advanced bonded airframe structure containing Ti-alloy, Al-50% B fibre composite and 6061 Al-alloy is shown in Fig 54.

10 TITANIUM SPF/DB STRUCTURES

The design and manufacture of SPF/DB structures is dealt with in other lectures in this series. However some brief comments on diffusion bonded joints in these structures are appropriate in this presentation. The formation of bonded joints prior to SPF has been described with reference to Figs 8-11 and the problems of contamination have been discussed in paras 4 and 5. A good example of diffusion bonds in a 4-sheet structure produced by British Aerospace is shown in Fig 55; no defects were detected in the db joints and full base metal properties should be realised for such joints in production[3]. Riveted and db joints are compared in Fig 56; much larger joint areas are obtained for db joints[3]. Typical shear strengths obtained for Ti-6Al-4V alloy are 10 MPa for riveted joints, 20-40 MPa for adhesive bonded joints and 575 MPa for db joints; the db joint therefore offers a significant static strength advantage[3].

However the stress concentration at the bond line between the box stiffeners and panel skin can be severe. Furthermore the edges of the bond have experienced shorter bonding times and consequently some initial fatigue crack growth is possible at these positions. Photoelastic measurements and fatigue tests indicate chemical milling and some redesign could lead to increased fatigue life for such joints[67]. In practice typical fatigue failure locations are associated with local sheet thinning at sharp radii rather than through the bond plane (Fig 56). When appropriate allowance is made for stress concentrations at notches and joints, current test data suggest that even bonded parts containing some microvoids will present no significant structural or design problem[3,4], although the overall acceptance requirements for SPF/DB structures in the presence of fatigue cracks and with few crack stoppers is still being considered[1].

Whilst the requirements for engine components differ in many respects from those for airframe components, the potential for SPF/DB components has been demonstrated by the hollow fan blade to be fitted to RB211-535 and V2500 engines[2,60]. The blade design requires up to 60° twist in the aerofoil, variations in sheet thickness and close tolerances to meet the operational load and foreign object damage requirements.

Notches associated with the bonds are also found in massive db components in Ti-6Al-4V, but these components have been accepted for service in aerospace applications on the basis of conventional inspection and test with reported fly-to-buy ratios reduced from about 8:1 to 3:1 compared with machined or forged components[44].

11 DETECTION OF DEFECTS IN DIFFUSION BONDED JOINTS

Apparently "perfect" diffusion bonded joints can have low strength[68] and adequate quality control and NDE is vital for economic SPF/DB processing[1]. The types of defect found in diffusion bonds have been described in sections 3-4 and classified for NDE purposes by Asarov[69] and Tober[68] in terms of their area and thickness (Fig 57). Ultrasonic techniques appear to have the greatest potential for detecting defects in diffusion bonds[62,70-72], but real time radiography offers the possibility of higher production rates with lower resolution eg ∿25 μm defects in 5 mm thick sections[73]. A high production rate is especially important for engine components[2].

Unfortunately the smallest defect size can be much less than the wavelength λ of conventional ultrasound (eg for a frequency of 30 MHz, $\lambda = 200$ μm), the reflection efficiency of defects is low and signal flight times of only 0.3 μs obtained with thin (1-2 mm) sheet SPF/DB structures requires high time resolution equipment. Consequently frequencies of > 30 MHz, point focussing probes and advanced data reduction techniques are required[68,72]. The minimum defect size detectable even under ideal conditions appears to be about 0.2-0.5 mm^2 in area and 1-3 μm thick[68,69]; some results obtained by Tober and Elze are given in Fig 58. Smaller microvoids and intimate contact disbonds cannot be detected by current NDE methods.

In practice therefore detection of microvoids depends upon destructive metallographic sectioning or proof testing[1]; sections are obtained either by cutting up selected production components or from cut-outs from components in the manufacturing stage. This quality control is backed-up by the use of correct processing parameters and by the imposition of close process control.

It is worth noting that for some bonds eg those in 3 or 4 sheet structures (Figs 22 and 23) the actual forming process effectively proof tests the bonded joints at elevated temperatures. Proof testing using acoustic emission has been used for brazed joints[74] and may be applicable to some db joints.

The importance of high quality in diffusion bonded joints should not be underestimated. At present two questions remain unanswered:-

i What is the maximum size/number of defects that can be tolerated in a diffusion bonded joint before the mechanical properties are adversely affected?

ii What is the minimum size/number of defects that are likely to escape detection in diffusion bonded joints made under production conditions?

If the size in (i) is less than the size in (ii) then diffusion bonded joints may be restricted to lightly loaded fail safe structures and the full potential of SPF/DB processing will not be realised.

12 CONCLUSIONS

Titanium alloy SPF/DB structures currently in production are a tribute to all those who studied and developed superplastic forming and diffusion bonding over many years. The emphasis in Ti-alloy SPF/DB structures is now moving towards process control and non-destructive examination to enable highly loaded structures to be designed for safe-life applications. Diffusion bonding for Al-alloys requires further work before high quality bonds can be reproduced and combined with superplastic forming. It is possible that for Al-alloys progress will be made in the future by exploiting experience gained in the manufacture of semiconductor materials to produce a surface coating pattern prior to bonding under automated environmental and process control. Massive db of smaller components in Ti-alloy is likely to increase in the future as the cost benefits that can be achieved in a dedicated processing plant are appreciated.

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N.B. An addendum follows at page 5-24.

TABLE 1

SOME CHARACTERISTICS OF DIFFUSION BONDED JOINTS

1	Joint strengths approaching or equal to the parent metal.
2	Bonding involves minimum distortion and deformation and close dimensional control is possible.
3	Large area bonds are possible, with improved joint efficiency compared with conventional joining.
4	Thick and thin sections can be joined to each other.
5	Cast, wrought and sintered powder products and dissimilar metals can be joined. May be only choice for metal matrix composites.
6	Process time independent of bond area or numbers of components.
7	Machining costs may be reduced.
8	Corrosion resistance of parent metal or of selected interlayer combination. No fluxes are required.
9	More efficient design and a smaller buy to fly rati [illegible] possible.
10	May be combined with superplastic forming.

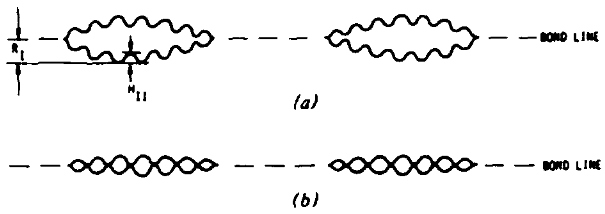
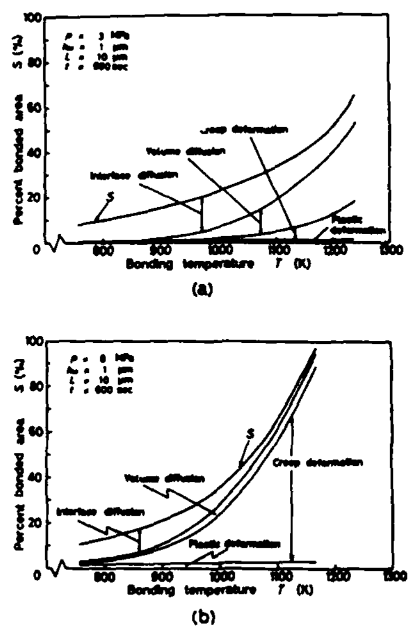


Fig 1 [Ref 10] -Schematic development of the bondline
The long-wavelength asperities are flattening during stage I in part (*a*). In part (*b*) the short-wavelength asperities remaining between the bonded regions have just achieved contact, and stage II closure begins.



Effect of bonding temperature on percent bonded area *S*, together with contributions of bonding mechanisms to *S*

Fig 2 [Ref 17]

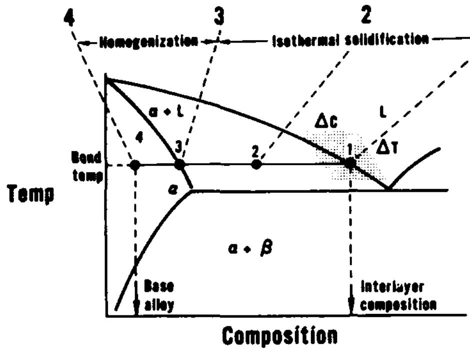
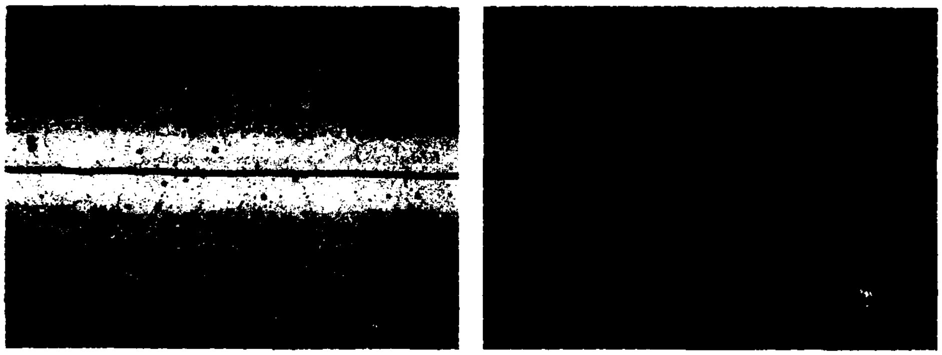


Fig 3 [Ref 23]



a. b.

Fig 4 Diffusion Bond Between Silver Coated Clad Al Alloy Sheet

a. 1 μm Diamond Polished Surface x 250 b. 600 Sic Ground Surface x 1000

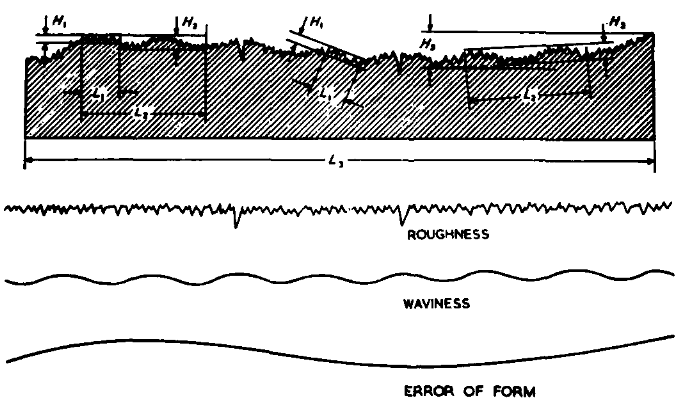
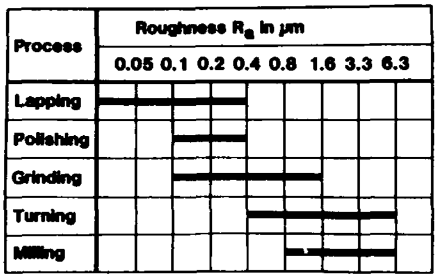
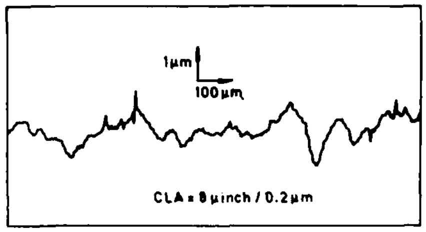


Fig 5 [Ref 24]

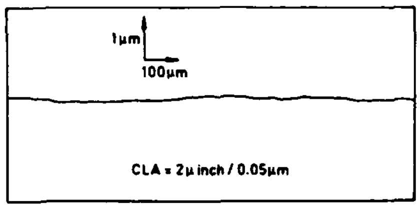


Typical roughness values obtainable by machining processes

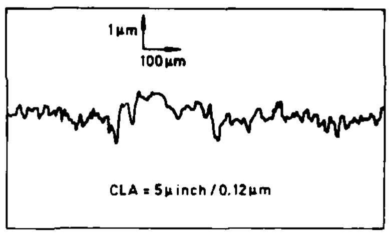
Fig 6 [Ref 3]



a. Talysurf traces for as received clad sheet



b Talysurf traces for 1 μm diamond polished clad sheet



c.

Fig 7 Talysurf Traces for Al Sheet

a. As Rolled b. 1 μm Diamond Polished
c. 600 SiC Ground

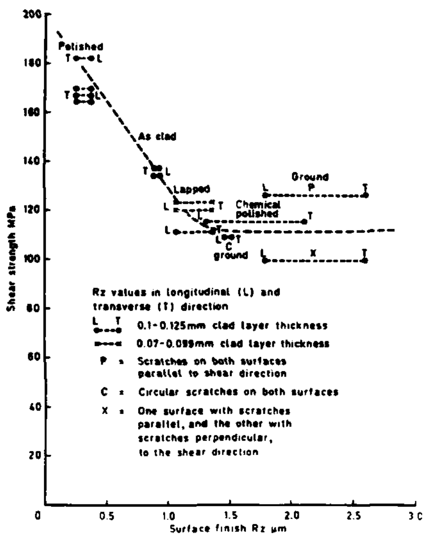
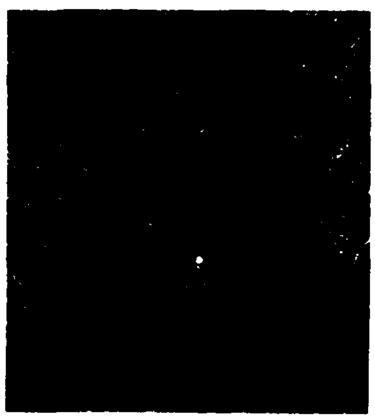


Fig 8 Shear Strength v Surface Finish for db Al-Alloy

a.

x 1000

b.



x 200

Fig 9 IMI 550 Alloy

a. As Ground Surface. b-d. Fracture Surface After db 4 min, 925°C, 0.52 MPa

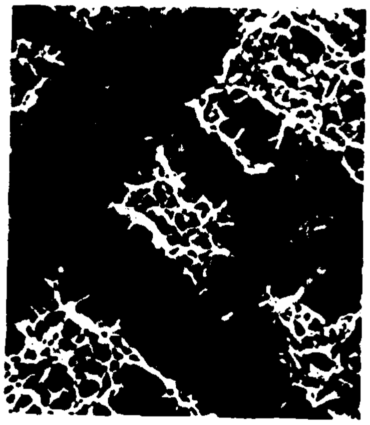
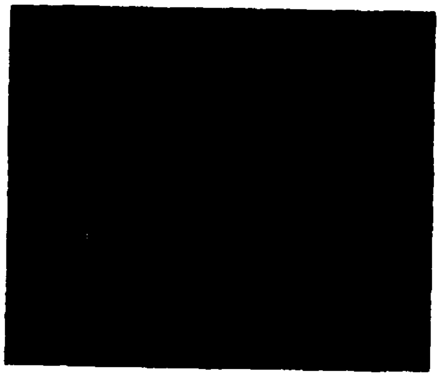
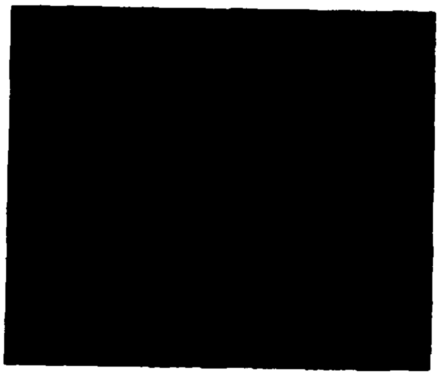


Fig 9 c.
x 1000

d.
x 1000



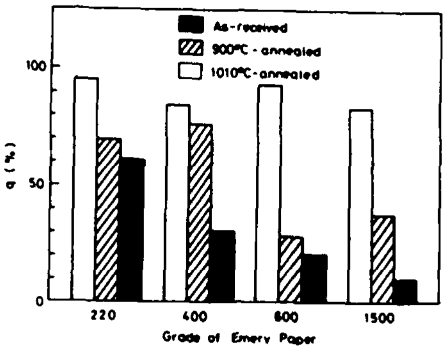
a. x 250



x 500
b.

Fig 10 Diffusion Bonded Joint After 0.5 h
925°C 0.52 MPa

a. IMI 550 b. IMI 318



Effect of faying surface roughness on the void ratio q for each base metal. T_W, P_W and t_W are 850°C, 0.2kg/mm^2 and 10min, respectively.

Fig 11 [Ref 28]

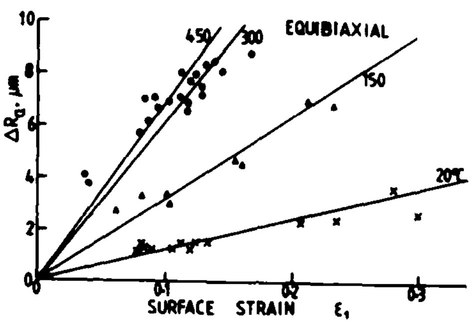


Fig 12 [Ref 30]

Surface roughening as a function of strain for material equibiaxially stretched at different temperatures.

Fig 13 Surface of Ti-6Al-4V Alloy After 300% Superplastic Strain at 875°C

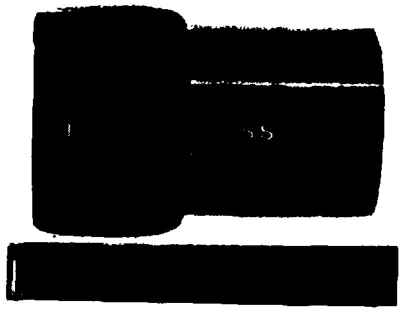


Fig 15 Ti-6Al-4V/Stainless Steel db Butt Joints

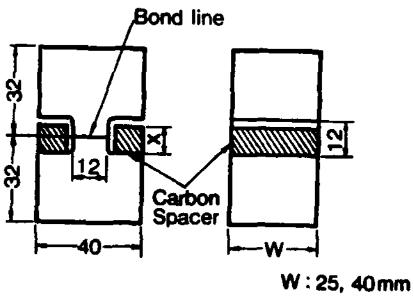
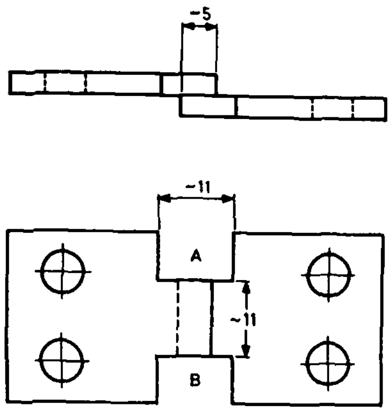


Fig 14 Specimen for Diffusion Bonding (W = width, X: spacer distance) [Ref 42]



Dimensions in mm

Fig 16 Overlap Shear Test Piece for Al-Alloys

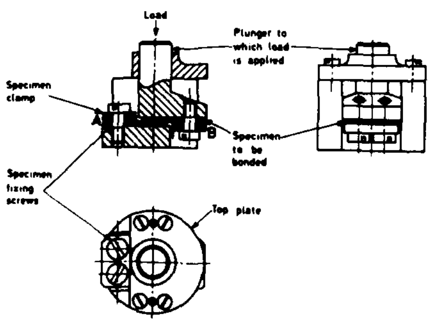


Fig 17 Diffusion Bonding Jig

a.

b.

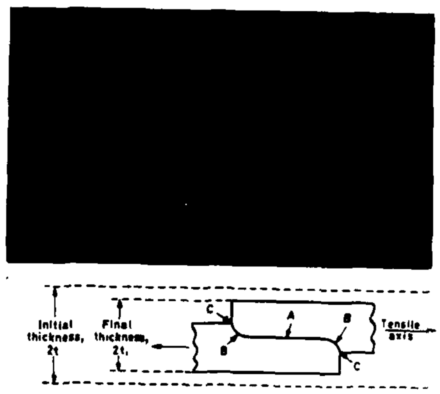
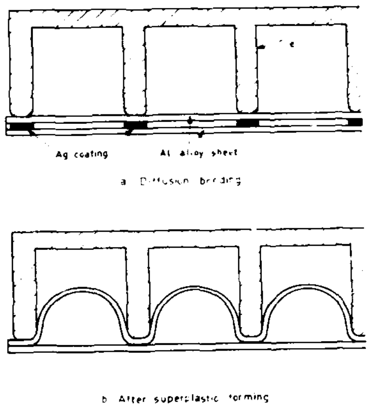
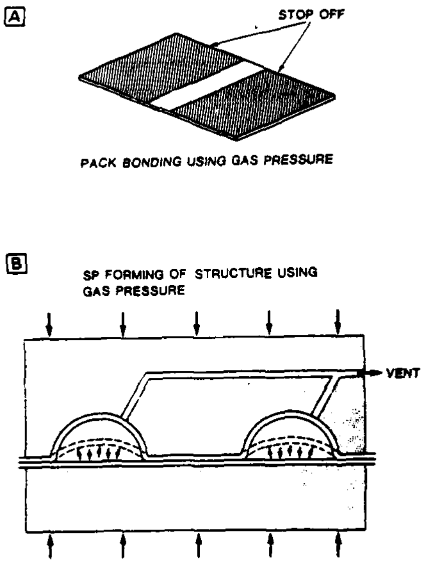


Fig 18 a. Section Through Lap Shear Test Piece
b. Schematic Diagram of Bond Interface

Fig 19 Diffusion Bond in SUPRAL 220 Alloy, Overlap <2t

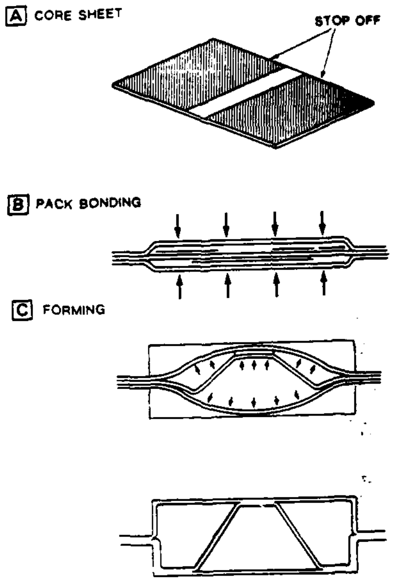
Fig 20





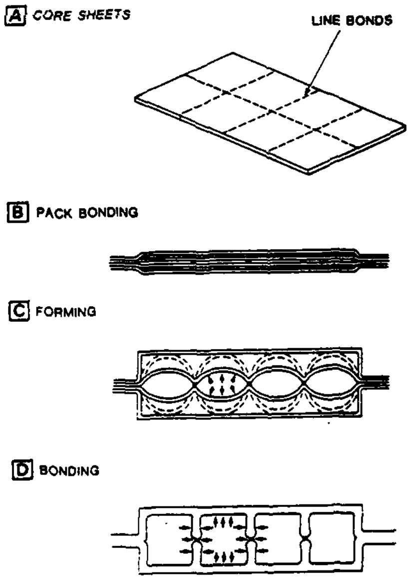
Two-sheet SPF/DB

Fig 21 [Ref 3]



Three-sheet SPF/DB

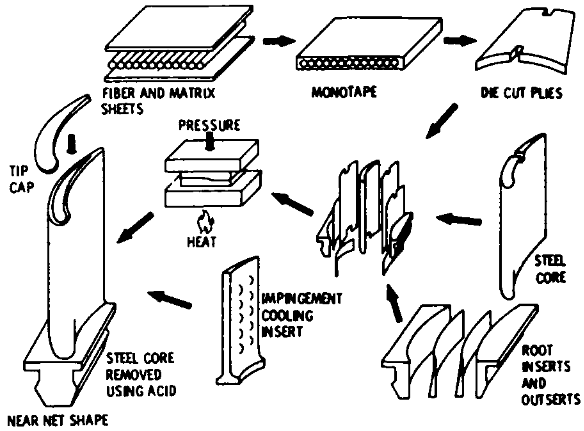
Fig 22 [Ref 3]



Four-sheet SPF/DB

Fig 23

Fig 24 [Ref 45]



TFRS blade fabrication process.

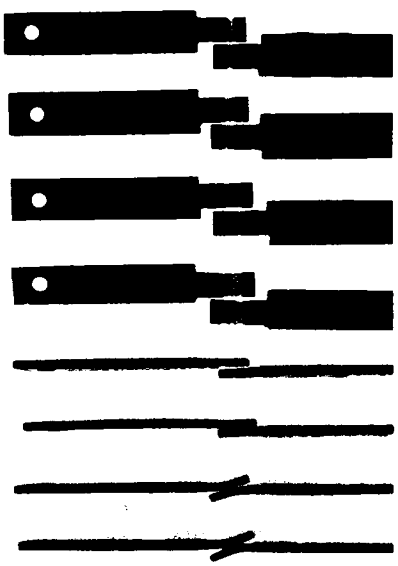
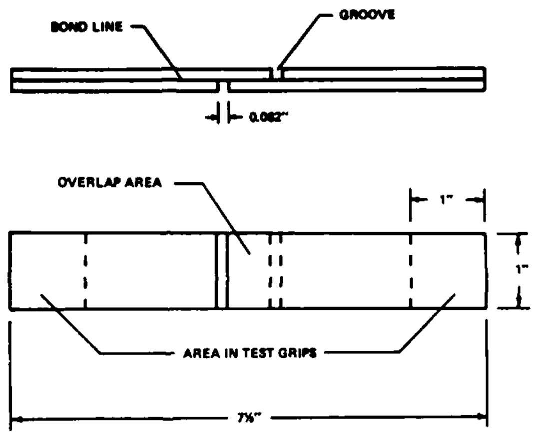


Fig 25 Effect of Increasing Overlap Length on Bending of db Test Piece x 0.55



LAP-SHEAR TEST SPECIMEN.

Fig 27 [Ref 49]

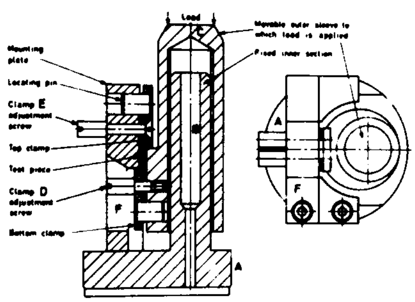


Fig 26 Lap Shear Test Jig

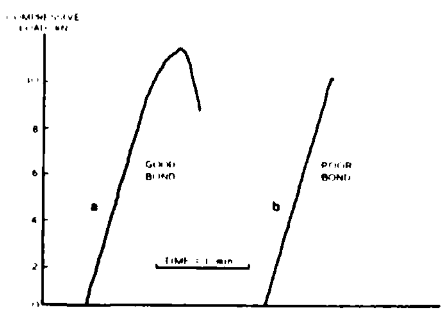


Fig 28 Load-Time Curves

a. Good Bond and b. Poor Bond in Shear
c. Good Bond in Peel

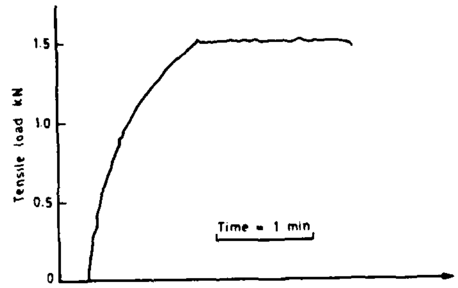


Fig 28 c.

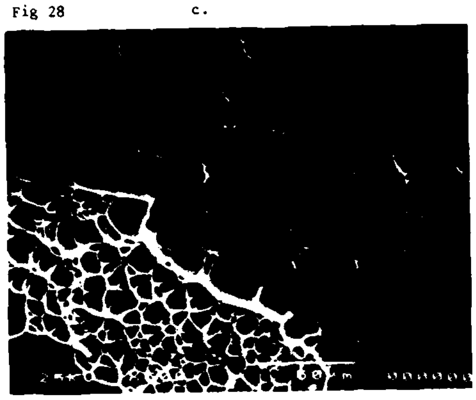
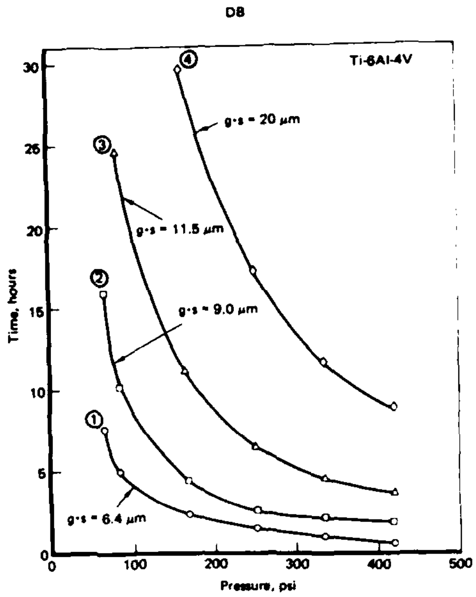
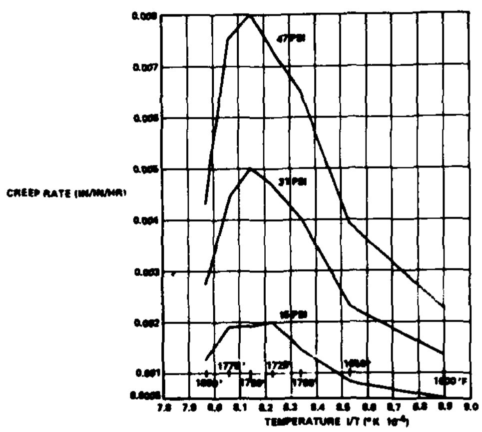


Fig 29 Diffusion Bond Shear Fracture Surface in Ti-6Al-4V x 500



Effect of grain growth on DB

Fig 31 Time v Pressure Curves for the Bonding of Ti-6Al-4V. [Ref 19]



Influence of temperature on the creep rate in compression at three different load levels

Fig 30 [Ref 52]

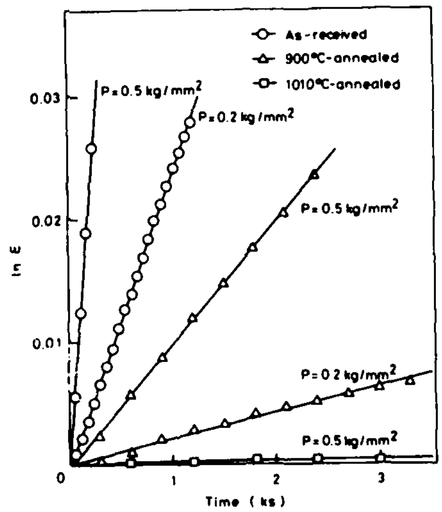


Fig 32 Semilogarithmic Plots of Strain ε against Time in a Compression Test at 850°C for each Base Metal. P Denotes the Compressive Stress. [Ref 28]

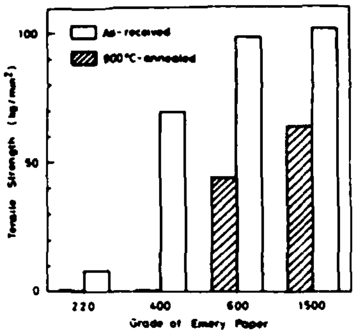


Fig 33
[Ref 28]

Effect of the faying surface roughness on the tensile strength of the joint of the as-received and 900°C-annealed base metal. T_w, P_w and t_w are 850°C, 0.2kg/mm^2 and 10 min, respectively.

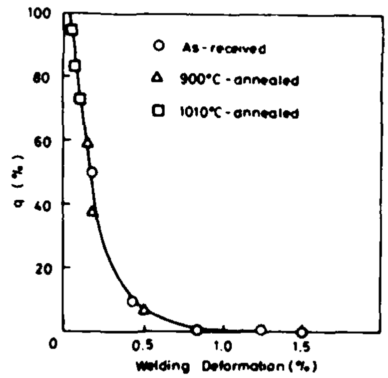


Fig 34
[Ref 28]

Void ratio q versus welding deformation for the joint of each base metal.

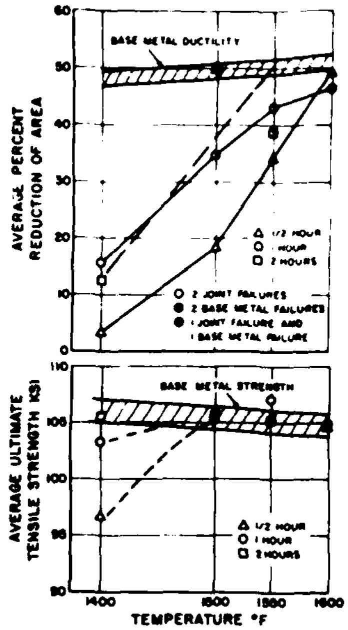


Fig 35 Plots of Average Joint Properties V Temperature for three times. [Ref 54]

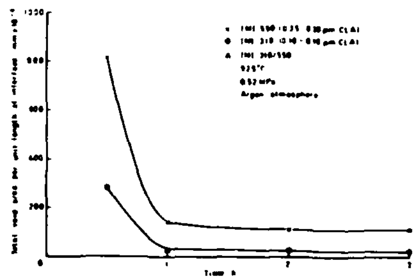


Fig 36 Effect of Bonding Time on the Void Content of Diffusion Bonded IMI 318 and IMI 550

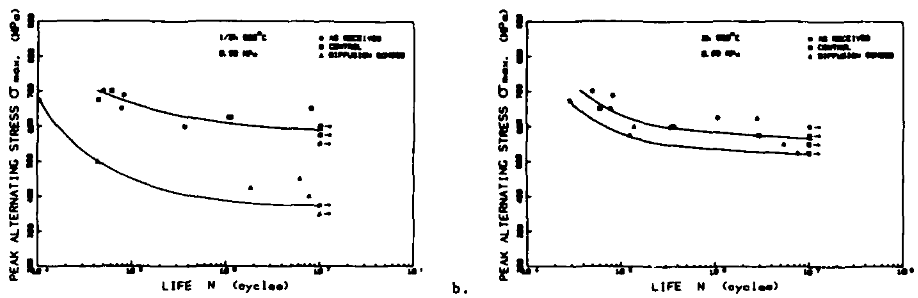


Fig 37 Peak Alternating Stress v Cycles to Failure for IMI 550 Alloy
a. db at 0.52 MPa, 0.5 h 925°C b. db at 0.69 MPa, 2h, 950°C

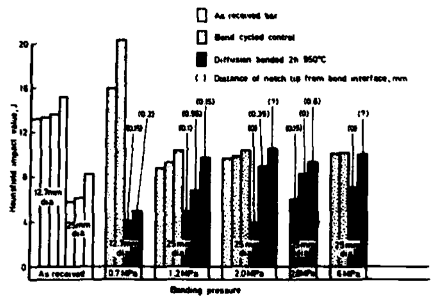


Fig 38 Impact Strength v Bonding Pressure for Ti-6Al-4V db Butt Joints

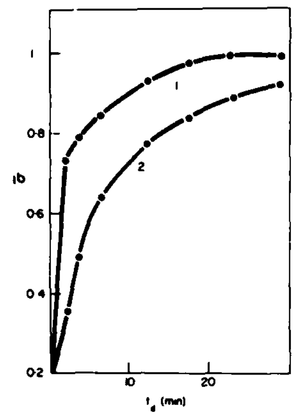


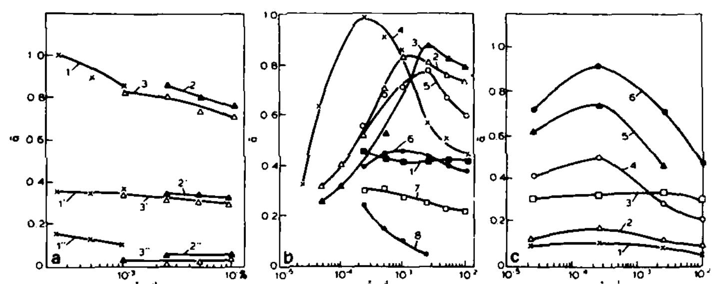
Fig 39 Relative Strength v Bonding Time

1) Pressure Cycled 0-0.9 σ_{ys}
2) Constant Pressure 0.9 σ_{ys}

[Ref 56]

Fig 40
Relative Impact Strength $\bar{a}$ v Strain Rate
Effect of
a. Bonding Deformation
b. Bonding Temperature for VT6 Alloy and
c. Effect of Bonding Temperature for Ni-alloy

[Ref 57]

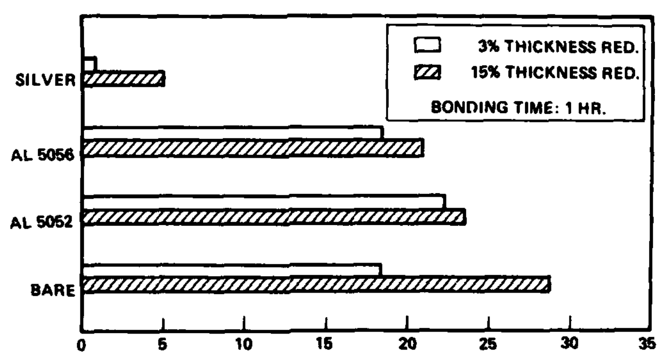


. A(ε) for BT-6 and nickel alloy welds. (a) BT-6: 1, 1', 1'', T = 960°C; 2, 2', 2'', 1,000°C; 3, 3', 3'', 1,040°C. 1, 2, 3, ε = 5%. 1', 2', 3', ε = 3%; 1'', 2'', 3'', ε = 1%. (b) BT-6: 1, T = 1,080°C. 2, 1,040°C; 3, 1,000°C; 4, 960°C; 5, 920°C; 6, 880°C. 7, 840°C; 8, 770°C. 1 - 8, ε = 5%. (c) nickel: 1, 5, T = 1,100°C; 2, 1,000°C. 3, 6, 900°C; 4, 800°C. 1 - 4, ε = 5%. 5, 6, ε = 10%.

EFFECTS OF INTERMEDIATE LAYERS AND AMOUNT OF DEFORMATION ON BOND STRENGTH

Fig 41
Bond Shear Strengths (psi x 1000) for 7475 Al-Alloy with Different Interlayers

[Ref 33]



BONDING TIME VS. BOND STRENGTH

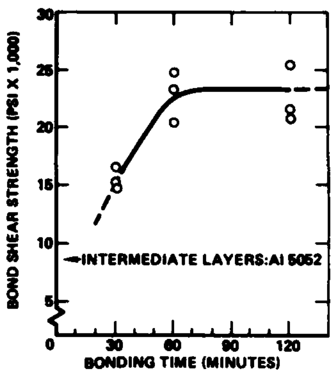
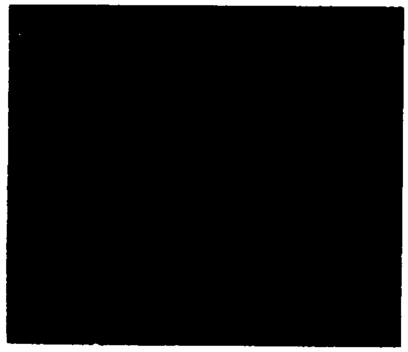


Fig 42 [Ref 33]

a.

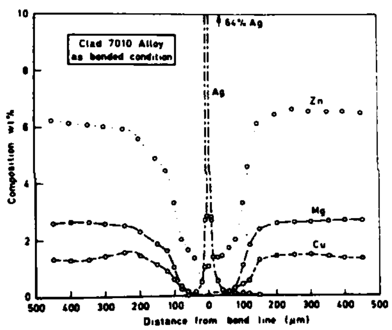


b.

Fig 43 Section Through db in Clad 7010 Al-Alloy

a. As Bonded Showing Silver Interlayer coating x 1110
b. After Solution Heat Treatment x 1110

a.



b.

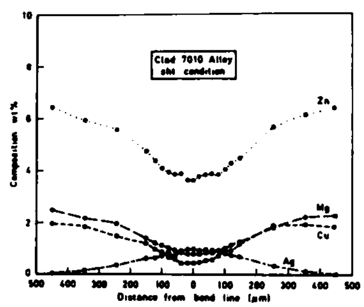
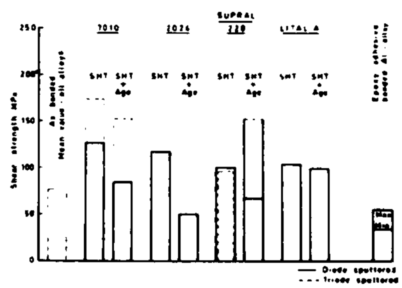
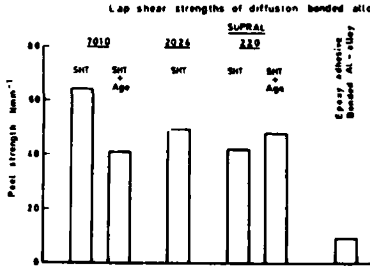


Fig 44 Composition Curves Across Diffusion Bonded Joint in 7010 Al-Alloy

a. As Bonded
b. After Solution Heat Treatment



Lap shear strengths of diffusion bonded alloys

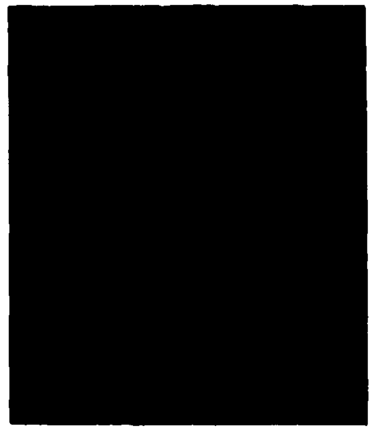


90° Peel strengths of diffusion bonded alloys

Fig 45

Fig 46 Plastic Deformation Associated with Crack Growth in Peel Test of Clad 7010 Al-Alloy Diffusion Bond. x 85

a.



b.

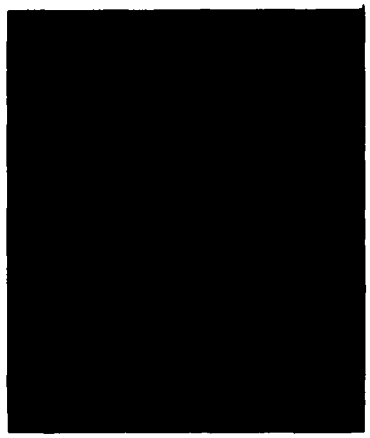


Fig 47 Shear Fracture Surfaces of Clad 7010 Al-Alloy Diffusion Bond

a. After SHT x 90   b. After Ageing x 90

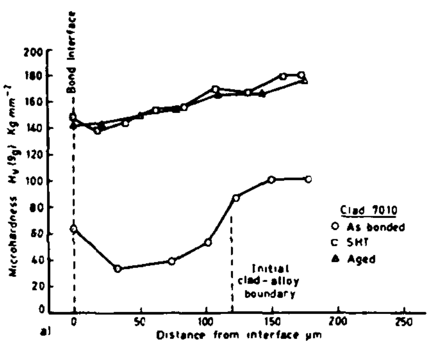


Fig 48 Increase in Hardness of Clad Layer in Bonded Clad 7010 Al-Alloy

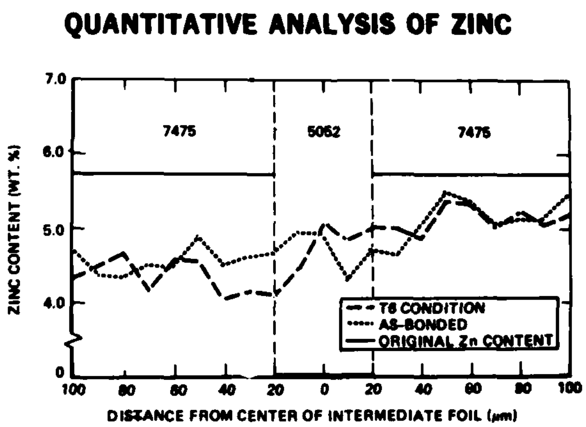


Fig 49 Diffusion of Zn During the Bonding of 7475 Al-Alloy. [Ref 33]

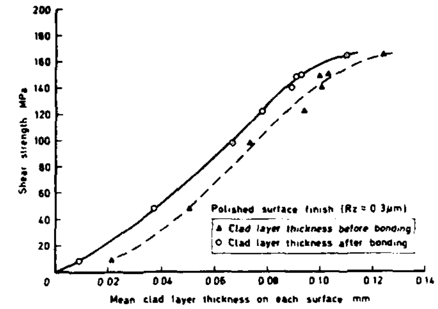


Fig 50 Shear Strength v Clad Layer Thickness for db 7010 Al-Alloy

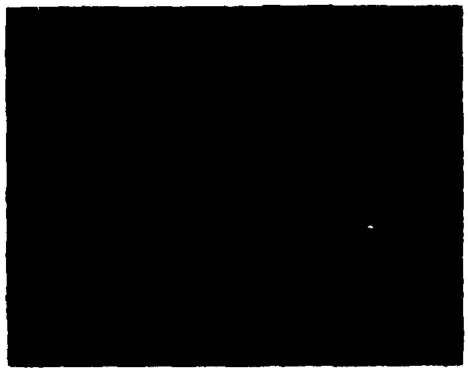


Fig 51 Intermetallic Layer at Bond Interface in Silver Coated Clad 7010 Al-Alloy after SHT x 1000

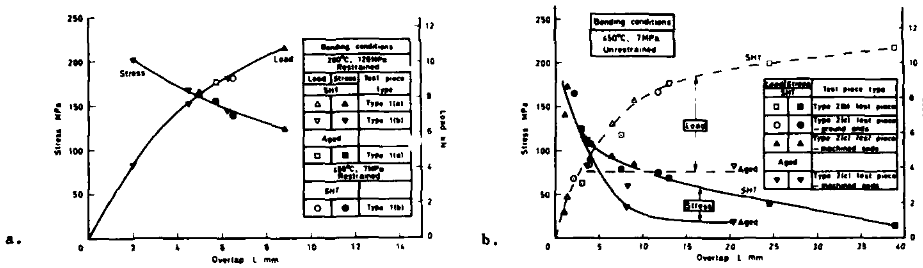
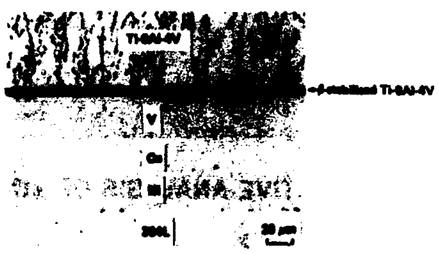


Fig 52 Effect of Overlap L on the Failure Load and Stress for

a. Type 1 Single Overlap Diffusion Bond

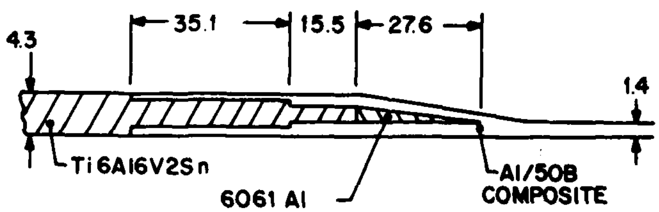
b. Type 2 Single Overlap Diffusion Bond Tested in the unrestrained state

Fig 53



Section of Bond No.2 (Ti-6Al-4V/V/Cu/Ni/304L)

Fig 54
[Ref 50]

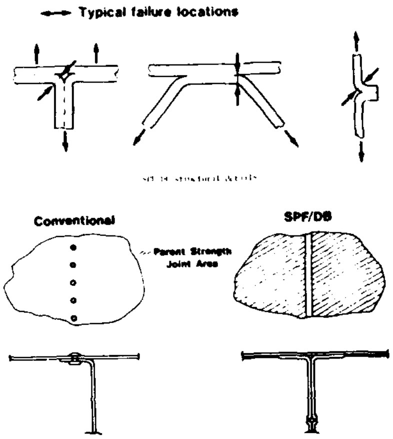


DIMENSIONS IN mm

—Diffusion welded joint of Al/B composite to Ti6Al6V2Sn

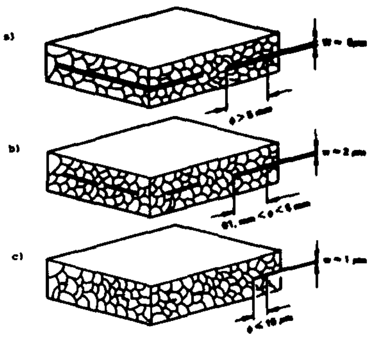
x 0.5

Fig 55 Diffusion Bonded 4 Sheet Structure in Ti-6Al-4V
Courtesy of British Aerospace (Warton)



Characteristics of SPF/DB structure

Fig 56 [Ref 3]



Classification of DB-defects

a) Coarse-dispersive macro defect
b) Fine-dispersive macro defect
c) Micro defect configuration

Fig 57 [Ref 68]

Fig 58 [Ref 68]

DB defects / US methods	Transmission non-focusing, focusing probes 5 and 10 MHz	Pulse/echo, non-focusing probes 10 up to 20 MHz	Pulse/echo non-focusing probes 20 up to 110 MHz	Pulse/echo point-focusing probes > 20 MHz
Course-dispersive macro defects (single defect size w ≈ 200 µm; φ ≈ 3 mm)	D	WD	WD	WD
Course-dispersive macro defects (single defect size w ≈ 5 µm; φ > 5 mm)	PD	D	WD	WD
Fine-dispersive macro defects (single defect size w ≈ 3 µm; 0,1 < φ < 5 mm)	PD	PD	WD	WD
Micro defect configurations (single defect size w ≈ 1 µm; φ < 15 µm)	ND	ND	PD	D

Detectability of DB-defects with different US-methods

ND not detected
PD partially detected
D detected
WD well detected

ADDENDUM TO 1987 AGARD LECTURE SERIES NO 154 ON DIFFUSION BONDING OF METALS

by
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A1 INTRODUCTION

This Addendum is intended to follow the 1987 Lecture on p 4-1 and to provide an update on developments since the last Lecture Series No 154 was published.

Recent international conferences and one text book are pertinent to work in this field[1-4]. The most successful large scale application of diffusion bonding is in the SPF/DB of thin titanium sheet aerospace structures[1-3] and this technology has developed into a competitive production process in many countries. Research continues on the development of a reliable cost effective SPF/DB process for Al-alloys with particular emphasis on lithium containing alloys in order to exploit the lower density and high specific stiffness of these alloys. The potential for metal matrix composites has directed attention not only to the mechanism of bonding the metal matrix, but also to the bonding of metal matrices to non-metallic barrier layer coatings on ceramic fibres. Dramatic progress has been made in producing ceramics with greater strength and toughness and this has led inevitably to increased effort on diffusion bonding of these materials to themselves and to metals. Diffusion bonding is therefore currently being considered for joining a broad range of materials in widely differing applications.

A2 DIFFUSION BONDING OF TITANIUM AND ALUMINIUM ALLOYS

In para 3 of the 1987 Lecture (p 4-2) the importance of surface roughness on the diffusion bonding parameters was emphasised. Recent work[5] has reported the increase in roughness caused by superplastic deformation (figs 1a-b). The roughness increased with increase in grain size and superplastic strain to Ra = 1.5 μm in Ti-6Al-4V alloy and to Ra = 2.7 μm in Al-Li alloy (LITALA). It is interesting to note that the specified surface finish for SPF Ti-6Al-4V sheet is Ra <1 μm[6]. The increase in roughness must therefore be taken into account when making secondary diffusion bonded joints in multiple sheet SPF/DB structures.

The emphasis in testing has been primarily on the behaviour of complete SPF/DB structures. The results indicate that the failure modes do not involve the db joint, but are confined to thin regions of the core sheets. Unfortunately it is impossible to carry out a shear test on a diffusion bonded joint between thin sheet, because of deformation and bending of the sheet. To overcome this problem a stack of sixteen 1.6 mm thick Ti-6Al-4V sheets has been gas pressure bonded to obtain large test pieces. Fatigue crack growth and fracture toughness data for specimens with notches in the bond plane indicated parent sheet properties were achieved[7]. However two Izod impact test pieces cut from adjacent regions of the stack had impact strengths of 12J and 22 J, as shown in fig 2a-b. The surface roughness of the fracture was greater for the higher toughness test piece and both fractures were parallel to the bond interface. These results confirm previous data which indicated impact strength of diffusion bonded joints was particularly sensitive to the quality of the diffusion bond interface. Nevertheless solid state and liquid phase diffusion bonded joints manufactured in Ti-6Al-4V alloy under carefully controlled conditions have been shown to exhibit parent metal impact and low cycle fatigue strength and have been successfully used in critical aeroengine components[8].

Conflicting strength data continues to be reported for diffusion bonded Al-alloy joints. The early work of Byun and Vatava (1987 Lecture Series p 4-5) reported shear strengths of 172 MPa and 241 MPa for 7475 alloy bonded with 5052 interlayer. In more recent work parent metal static and cyclic shear strengths were obtained without any interlayer or coating[9]. However other workers have reported[10] much lower strength (<80 MPa) with a variation as high as ±33 MPa which led to data being analysed using the Weibull probability function. Such variability would be unacceptable in a manufacturing environment and emphasises the need to identify and control the factors giving rise to bond strength variability in Al-alloys. The most important factor is probably surface contamination, particularly by oxide films. In work on bonding Al-Li alloys in vacuum, the less protective oxide film on these alloys enabled parent metal solid state bonds to be made more easily contrary to the conclusion that lithium additions made bonding more difficult[11]. A significant factor controlling bond strength in superplastic 8090 Al-Li alloy appeared to be the planar bond interface (fig 3). Resistance to grain boundary migration is desirable for superplastic forming, but is a disadvantage for diffusion bonding. Similar strength bonds can be obtained by transient liquid phase bonding with copper interlayers, but the copper concentration must be controlled to avoid residual eutectic phase at the bond interface.

A3 DIFFUSION BONDING OF DISSIMILAR METALS

Dissimilar metal joints may be used to:

a. Conserve strategic or scarce metals.

b. Provide oxidation, corrosion or wear resistance without sacrificing mechanical properties.

c. Tailor microstructure - in a particular part of a component.

Diffusion bonding techniques are therefore being developed for components with a wide range of alloy combinations and diffusion bonding may be combined with a hot isostatic pressing operation[12,13]. The components are then evacuated and sealed in a can before isostatic pressing. This avoids gas entrapment at bond interfaces (particularly important when multiple interlayers are used), inhibits Kirkendall void formation and allows complex interface contours to be manufactured.

Only limited studies have been carried out on the strength of dissimilar metal joints and the many variables involved are often interdependent, eg the thickness, composition and strength of the interlayers may be dictated by the need to avoid intermetallic formation or high thermal stresses. The strength is dependent on the fracture path, which is shown schematically for a Ti-6Al-4V/V/Cu/Ni/stainless steel bond in fig 4[7]. Although fracture predominated in the Cu/Ni interface, fractures in both Cu/V and Ni/stainless steel were sometimes observed. An understanding of the microfractography in these types of joint is therefore necessary if optimum joint mechanical properties are to be achieved. In practice fracture strength is usually limited by the weakest interface or interlayer. This limitation may be reduced or avoided in shear loaded joints by producing a non-planar interface as shown in fig 5; the shear strength is then related to the bulk shear strength of the weakest of the two alloys being joined.

Roll bonding is a well known technique for cladding sheet, but requires large reductions in thickness. Interesting developments in vacuum rolling (2×10^{-5} torr) in the USSR and USA[14] has enabled tne bonding of thin sheets (~0.13 mm thick) to dissimilar metals with about 10% reductions in thickness at relatively low temperatures which avoids the formation of intermetallics, eg for bonding to iron-base alloys the temperatures were 650°C (Ni), 800°C (Monel, Type 304 stainless steel, titanium) and 1100°C (Nb). In a similar development a metallic glass was roll bonded to nickel coated iron without recrystallising the glassy phase.

A4 DIFFUSION BONDS IN SHORT OR CONTINUOUS FIBRE METAL MATRIX COMPOSITES

The manufacture of continuous fibre reinforced sheet usually involves diffusion bonding in an evacuated and sealed container of either a sandwich of metal sheet and fibres or of filament wound coated fibres[15]. The most common fibre is SiC, although alumina, boron and graphite fibres are also used[16] Melt infiltration or squeeze casting may be used to produce an Al-alloy matrix, but molten metal techniques are not considered suitable for high melting point matrices such as titanium alloy and superalloy. In the manufacture of a Ti-alloy metal matrix composite (MMC) the bonding of the matrix to the fibre must be considered. Aerospace companies receive MMCs in the form of monoply or multiply (ie with one or more fibre layers) laminates which are then diffusion bonded to:-

a. Each other to increase thickness or to make close-to-form parts.

b. Conventional thin sheet as "patches" to increase local stiffness and strength.

In both (a) and (b) diffusion bonding is more difficult than with a fibre free matrix because the surface is less flat (thickness of Ti-alloy MMC monoply can vary by 40%) and where fibres cross each other it is more difficult to obtain surface contact. The type of defect that can arise in Ti-6Al-4V/SiC MMC is shown in fig 6; these defects may be impossible to detect by NDE techniques (see 1957 Lecture Series p 4-6).

Diffusion bonded joints with bond interfaces normal to the fibre direction have been produced in Al-Mg-Si/short Al_2O_3 fibre composites[17]. When particular care was taken in the surface preparation to avoid protrusion of the fibres, intimate surface contact was obtained (fig 7b) and bond strengths equal to that of the base metal were possible. However protruded fibres prevented diffusion bonding (fig 7c).

The mechanical properties of MMCs are likely to be limited by the strength of the fibre/matrix interface. Reactions at this interface depend upon the system[16,18], eg in Al-alloy matrices B-fibres react at 400°C and Al_2O_3, SiC and C-fibres react at 500°C. The reaction is more severe in the Ti-alloy/SiC fibre system at 900°C, but a 3 µm thick C-rich coating significantly reduces the rate of attack[19]; B, C and SiC fibres dissolve in superalloy matrices at high temperatures. Thus diffusion bonding of MMCs is a complex process requiring only sufficient diffusion across metal/ceramic and possibly ceramic/ceramic interfaces to strengthen, but not degrade the interfaces. Unfortunately it is very difficult to isolate the contribution of fibre/matrix interfaces to the strength of an MMC containing hundreds of fibres, some of which may become fractured during processing or during testing. Progress in optimising the mechanical properties of MMCs in the future will probably be made by studies on single fibres embedded in a matrix[19].

A5 DIFFUSION BONDING OF METALS TO CERAMICS AND CERAMICS TO CERAMICS

Ceramics (Al_2O_3, SiC, Si_3N_4) are being studied extensively in many countries because of their excellent corrosion and wear resistance, high temperature properties, oxidation resistance and low densities. However their disadvantages are lack of ductility, sensitivity to defects and large variability. A possible compromise is ceramic bonded to metal, but ultimately monolithic ceramics are required with improved toughness. This might be achieved by producing ceramic matrix/ceramic fibre composites.

An increasing number of reports are being published on the joining of ceramics to themselves and to metals. Diffusion bonded joints appear to have higher strengths than brazed or mechanical joints[20,21]. However covalent bonding in ceramics makes them difficult to join to metals and interface contact must be achieved either by plastic deformation in the solid state of the metal (in ceramic/metal joints) or of soft interlayers or liquid phase brazing must be used.

Ti-alloys reduce most ceramics in the solid state. In contact with Al_2O_3 for example in vacuum at 945° for 2h a 2 µm thick reaction layer occurs. Bonding at 1500°C produces a 70 µm thick layer, multiple cooling cracks and hardness values in the layer of up to 1000 HV[22]. High bonding temperatures (~1000°C) accentuate the thermal stress problems in bonded ceramic joints, but these can be significantly reduced by the use of multiple interlayers. This has enabled high bond strengths to be obtained between Nimonic 80A and SiC or Si_3N_4 in the solid state[21]. Interfaces of the type Si_3N_4/Ni(0.5)/W(0.8)/Cu(0.5)/Nim 80A where the numbers denote interlayer thicknesses, in mm were bonded at 950°C for 2h at 55 MPa pressure to produce tensile strengths of ~124 MPa. High bonding pressures are avoided by using braze metal interlayers. For example joints made between Al_2O_3 at 1100°C using an Al-4% Cu braze metal had shear strengths of 157 MPa[23].

The origin of the strength in brazed ceramic/ceramic joints is not always clear. However careful studies carried out on SiC joined by Ag-Cu-Ti braze metal at 950°C for 30 minutes revealed a remarkable match between the lattice planes in SiC and lattice planes in a TiC interaction layer. The high bending strength of >300 MPa was attributed to this excellent matching between braze and ceramic[24]. Multiple interlayers have also been used to join Si_3N_4 in the solid state in vacuum at 950°C under a pressure of 20 MPa[25]. In this case it was found that interdiffusion in the interlayers activated the diffusion bonding process. High bend strengths up to 290 MPa were obtained with interfaces of the type Si_3N_4/Ni(0.01)/Nb(1.0) Ni(0.01)/Si_3N_4.

One of the most exciting developments in ceramics has been the observation that Yttria stabilised zirconia (TZP) and TZP + Al_2O_3 composites can be made superplastic. This has direct implications for diffusion bonding, since good interface contact can be achieved under superplastic conditions. Consequently diffusion bonded joints were produced in these ceramics at 1475°C under a pressure of 12.5 MPa to give a bonding strength of 1360 MPa[26].

A6 CONCLUSIONS

The trend is towards joining advanced materials by techniques based upon diffusion bonding in order to obtain high strength joints. Superplastic deformation and diffusion bonding has now be combined to join both metallic and non-metallic materials. Further progress will depend upon a better understanding of the kinetics and strengthening mechanisms in joints between an increasing range of materials.

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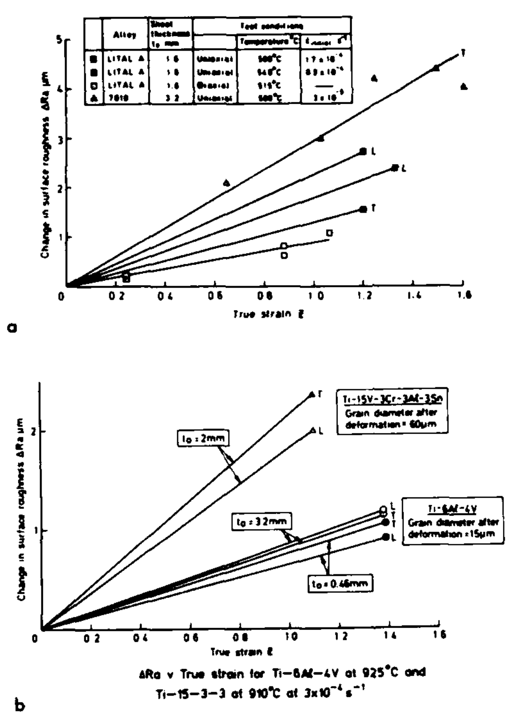

Fig 1 Change in surface roughness (ΔR_a) versus true strain for alloys deformed under superplastic conditions.

a. Al-Li (8090) alloy, initial and final grain sizes 5 μm and 12 μm respectively. Al-Zn-Mg (7010) alloy, corresponding grain sizes 8-12 μm and 12-24 μm.

b. Ti-15V-3Cr-3Al-3Sn and Ti-6Al-4V alloys.

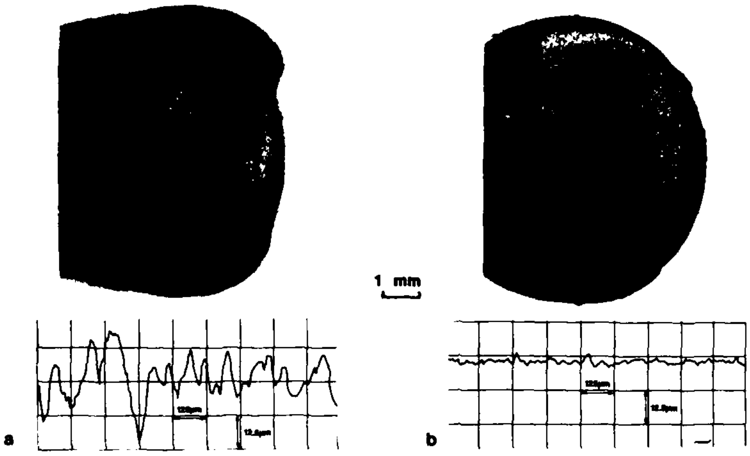

Fig 2 Izod impact fractures (x8.6) and fracture surface roughness for test pieces machined from a diffusion bonded stack of 16 Ti-6Al-4V sheets.

a. Impact strength 22J.

b. Impact strength 12J.

Fig 3 Section through a diffusion bonded joint in Al-Li (8090) alloy after SHT + age.
Bond interface A-A.

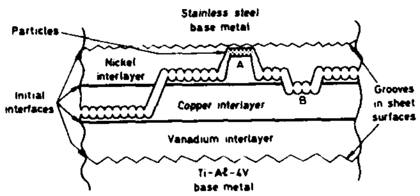

Fig 4 Schematic diagram of shear fracture planes in diffusion bonded joint between Ti-6Al-4V and S526 stainless steel with V/Cu/Ni interlayers.

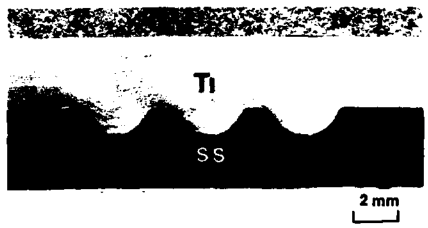

Fig 5 Non-planar interface produced in a Ti-6Al-4V/stainless steel joint. Ti = Ti-6Al-4V SS = S526 stainless steel.

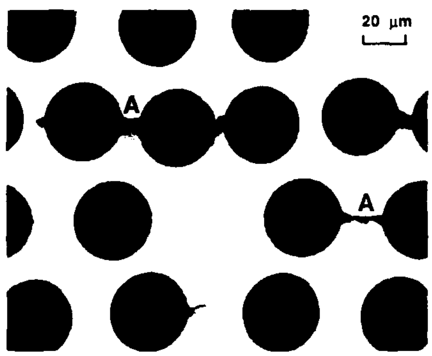

Fig 6 SiC fibres sandwiched between Ti-6Al-4V foil and hot vacuum press bonded. Unbonded areas at A.

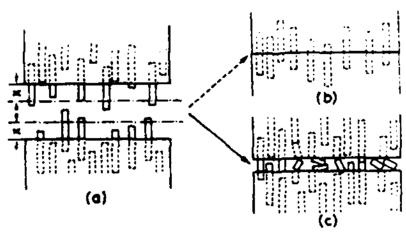

Fig 7 **Schematic diagram illustrating the behavior of protruded fibers: (a) before contact, (b) ideal contact and (c) real contact.**

THE MECHANICAL PROPERTIES OF SUPERPLASTICALLY FORMED TITANIUM AND ALUMINIUM ALLOYS.

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SUMMARY

The behaviour of aluminium and titanium alloys under superplastic forming conditions is well documented but there is much less published data on the effect of the superplastic forming process on the mechanical properties; these data are essential for the design of structures. This paper reviews the effect of superplastic forming parameters such as temperature, strain rate, strain and post forming heat treatments upon the tensile, fatigue and fatigue crack growth performance of these alloys and relates the property variation to changes in the microstructure. During superplastic forming of aluminium alloys intergranular cavities are formed with increasing strain which degrade the material and reduce the mechanical properties. Ways to prevent cavitation both during and after superplastic forming have been developed and the effect of these treatments on the mechanical properties will be discussed.

1. INTRODUCTION

Superplasticity is defined as the ability of a material to undergo large elongations, of the order of several hundred percent, without 'necking' and failure. The predominant deformation mechanism is grain boundary sliding, which is favoured by high temperatures (>0.5Tm, where Tm is the melting point in degrees absolute) and controlled strain rates. An ultra-fine grain micro structure (<10μm) is required which is relatively stable at the high forming temperatures.

Superplastic forming (SPF) offers considerable potential for reducing the weight and manufacturing cost of typical aircraft structures, particularly if it is combined with diffusion bonding (DB). As a result, considerable effort has been devoted in the aerospace industry to the development of SPF techniques for both titanium and aluminium alloys. In order to fully exploit the process, it is necessary to be aware of the potential for novel designs and to have mechanical property data for the formed material. Possible failure modes also need to be understood in SPF/DB structures.

In spite of the extensive literature which is available on the superplasticity of metals, there is relatively little information on post-forming mechanical properties. Initially, this was probably due to the lack of processing plant capable of producing large areas of superplastically formed material. Problems were also encountered in the extraction of suitable test specimens from formed shapes, due to significant thickness variations over the surface area. Most of the early data was therefore generated from uniaxially deformed bar or sheet, processed in tensile test machines modified to operate at constant strain rates and elevated temperature.

2. TITANIUM ALLOYS

The mechanical property data that is presented for superplastically formed titanium alloys has been obtained from material deformed using one of the three techniques described below:-

i) UNIAXIAL FORMING, where the material is subjected to a uniaxial tensile force at elevated temperature and in an inert argon atmosphere.

ii) MUFFLEBOX OR COLD DIE TECHNIQUE, where the material is loaded into a cold die, sealed and heated to temperature. It is then formed using an inert gas and cooled to room temperature before being removed from the tool.

iii) HOT OPEN DIE TECHNIQUE, where the material is loaded into hot dies (i.e. at working temperature), formed using an inert gas and removed to cool in still air.

2.1 Microstructural Effects.

Ti/6Al/4V has a two phase $\alpha+\beta$ microstructure and is the most widely used alloy for producin SPF and SPF/DB structures. Optimum forming parameters depend on the initial grain size and volume fractions of the alpha and beta phases that are present. Initially, forming temperatures of around 925°C and strain rates of the order of 1.5 x 10-4 sec-1 were used [1,2,3], but today, the trend is towards much lower temperatures and higher strain rates in order to improve efficiency. Even with these improved conditions, the material is still held at high temperatures for long periods of time (1-4 hours), which would allow recrystallisation and grain growth to occur.

The effect of the thermal cycle on grain size can be determined from a series of isothermal anneals, and typical results for water quenched Ti/6Al/4V are shown in Figure 1a. For isothermal anneals, the actual grain growth is found to be negligible, but the combination of temperature and superplastic strain produced a significant increase in grain size. However, not all Ti-alloys behave in a similar manner. Isothermal grain growth is much more rapid in beta stabilised Ti/15V/3Cr/3Al/3Sn, but superplastic deformation ultimately results in an overall reduction in grain size (4) (Figure 1b).

2.1 (contd......)

The post-formed microstructure is also dependent on the forming technique and typical microstructures for Ti/6Al/4V in the as-received and formed condition are shown in Figure 2. Changes to the grain size, shape and beta distribution after forming are clearly visible. The difference between the microstructures for the mufflebox and hot open die technique are due to the effect of cooling rate. With the mufflebox process, the formed part is removed from the furnace in the tool and slowly cooled to room temperature before removal. Parts formed using the hot open die technique are extracted at forming temperature and allowed to cool quickly in still air. Minor variations in microstructure can result from different cooling rates associated with differences in final thickness or component configuration with SPF/DB parts. These differences in microstructure account for the differences in mechanical properties between the two processes.

Superplasticity is characterised by a resistance to necking during deformation, which enables large strains (300-500%) to be achieved in sheet whilst retaining uniform thickness. Any local variations in superplasticity will lead to local necking and variations in the final thickness. Such an effect has been observed in Ti/6Al/4V when continuous bands of the alpha phase are present in the original sheet microstructure (Figure 2b).

Banded microstructures tend to be more pronounced in the bar form, and aligned alpha bands have been shown to resist deformation and give rise to anisotropic stresses and strains [5] The resulting shape changes in round bars after superplastic deformation are shown in Figure 3a. The peaks on the surface correspond to the ends of the α-bands which are perpendicular to the deformation axis. A similar effect has been observed in thin sheet (Figure 3b) which has been machined from bar [4]. During superplastic deformation, significant reductions in the amount of banding and near isotropic deformation was obtained after 60% uniaxial strain. The banded structure was fully eliminated after 300% strain giving a more homogenous microstructure which then deformed isotropically.

These results clearly show the importance of thermo-mechanical processing in the production of superplastic quality sheet. It is not sufficient to produce a small grain size, a uniform distribution of the phases is also need to give consistent parts.

As stated in the introduction, any reduction in forming temperature will reduce energy and tooling costs. However, reducing forming temperature not only reduces superplasticity, it also leads to higher forming stresses as the amount of beta in the microstructure decreases. Slip modes also start to operate in addition to grain boundary sliding, and the deformation approaches that characteristic of conventional hot forming [6]. The ability of the material to change shape then becomes more dependent on crystallographic texture, and there is a greater tendency for variations in sheet thickness and for anisotropy to occur. Attempts at reducing the forming time by increasing the strain rate have a similar effect.

It should also be noted that titanium has a high reactivity with a number of elements (oxygen, hydrogen, nitrogen, etc), which diffuse into the material and produce a hard, contaminated surface layer. As with the general microstructure, the type and degree of contamination is dependent on the forming technique that is used. With the mufflebox process the titanium material is heated, formed and cooled in an inert atmosphere, minimising the opportunity for contamination to occur. When present, it can take the form of an alpha or beta stabilised layer, depending on the source and hence type of foreign atom present. In the hot open die process, the material is exposed to the atmosphere during both the heating and cooling cycles. As a result, the formed parts usually have an alpha contaminated layer and the degree of contamination is generally greater than that produced by the mufflebox process.

Contamination is always removed from the surface of formed components by a subsequent chemi-milling operation, and all of the mechanical properties that are quoted are for material with no contamination present.

2.2 Uniaxial Tensile Property Data

The effect of superplastic strain on the strength of Ti/6Al/4V sheet material is shown in Figure 4. The actual strength values depend on the original sheet thickness and processing history [7], but the trends are similar for each of the starting gauges. Basically, a significant reduction in strength is caused by the thermal cycle, with a further and much smaller reduction in strength resulting from the superplastic strain.

IMI550 has a similar superplastic behaviour to Ti/6Al/4V, but at a lower optimum forming temperature of 900°C. Overall trends (Figure 5) are very similar to Ti/6Al/4V. However, the alloy is heat treatable and final strength can be influenced by the subsequent thermal history. Room temperature tensile properties for IMI550 material in a number of heat treatment conditions and in the unformed and superplastically formed state are summarised in Tables 1 and 2. From the quoted figures [7, 23, 24] it is apparent that $\sim$5% increase in tensile strength can be achieved by increasing the cooling rate after forming from 25°C min^{-1} to 150°C min-1, with a further increase of $\sim$ 8% by subsequent ageing. A cooling rate of 150°C min-1 after SPF and an ageing treatment of 500°C for 24 hours produced a tensile strength which was 4% higher than the mill annealed material and considerably greater than Ti/6Al/4V material in the as formed condition.

2.2 (contd......)

The Ti 8.1.1 and Ti 15.3.3.3 alloys have also been evaluated [7] but the results were disappointing. Ti 8.1.1 only exhibited high strain rate sensitivity ('m' value) at high temperatures (940-1010°C), whilst Ti 15.3.3.3 had low 'm' values and a coarse grain size which led to severe surface rumpling. Corresponding tensile strengths after SPF were also much lower than for Ti/6Al/4V.

2.3 <u>Biaxial Property Data</u>

A number of SPF boxes have been produced from nominal 3mm thick Ti/6Al/4V sheet using the mufflebox [8] and Hot Open Die [9] techniques. A series of test specimens (tensile, fatigue, crack propagation/fracture toughness) were then extracted from various positions in the bases to give longitudinal and transverse (to final rolling direction) properties for a range of superplastic strains. Tensile and fatigue results are presented in Figures 6,7 and 8.

2.3.1 <u>Tensile</u>

The mufflebox process produced similar results to the uniaxial data. Thermal cycling alone accounted for $\sim$ 10% reduction in strength, with a further 2-5% reduction being attributable to superplastic deformation. Overall trends were the same for the hot open die technique but to a lesser extent. In general, these differences can be explained by the different microstructures resulting from different cooling rates. The only anomaly is an apparent increase in tensile strength with increasing strain for the hot open die technique. This effect may be due to diffusion of the contaminating interstitial elements (oxygen, hydrogen, nitrogen, etc) from the surface into the bulk material.

2.3.2 <u>Fatigue</u>

The notch fatigue performance of Ti/6Al/4V material deformed using the mufflebox process has been evaluated using constant amplitude loading at a stress ratio R of 0.05 and a Kt of 2.68. The nominal starting thickness in this case was 2.0mm and it was subjected to superplastic strains of up to 150%. From the resulting S/N curve (Figure 8a) and a limited number of as-received and heat cycled specimens, it was evident that superplastic strain had a negligible effect on fatigue performance.

A comparison between the fatigue performance of mufflebox and hot open die material under FALSTAFF flight by flight loading is given in Figures 8b and 8c. The specimens were notched to give a stress concentration factor of Kt=2.44 and tested at two different load levels (σ 32= 490MPa and σ 3' [illegible] a[illegible]. [illegible]though the results exhibited a relatively large amount of scatter, the HOD technique generally produced slightly better lives. This is not surprising given the slightly better static properties for the HOD technique

2.3.3 <u>Crack Propagation/Fracture Toughness</u>

A comparison of crack propagation characteristics and fracture toughness results for as-received and SPF material showed that superplastic deformation did not have a detrimental effect on properties.

<u>ALUMINIUM ALLOYS</u>

As with titanium alloys, the major pre-requisite for superplastic deformation is a fine, stable grain size. However, unlike the duplex titanium alloys, the aluminium alloys generally have a single phase microstructure which tends to lead to grain growth and intergranular cavitation during high temperature forming.

The aluminium alloys which are currently available as production quality SPF materials are based on the Al/Zn/Mg and Al/Cu/Zr systems. In the case of the Al/Zn/Mg material (7475) superplasticity has been achieved by special thermomechanical processing in order to give fine recrystallised grains which are controlled by a dispersoid particle distribution. The Al/Cu/Zr family of alloys (Suprals) rely on dynamic recrystallisation during the forming process to produce and maintain the fine grain size.

During development of the Al/Li/Cu/Mg/Zr alloys (Alcan 8090 and 8091), it was found that these materials were also capable of superplastic deformation. The initial fine grained unrecrystallised microstructure of the as received material is converted to a fine grained recrystallised structure during superplastic strain.

3.1 <u>Microstructural Effects</u>

With superplastic deformation of aluminium alloys, the absence of an additional phase which is capable of accommodating grain boundary sliding leads to the formation of intergranular cavities. The effect of superplastic forming strain on microstructure is clearly shown in Figure 9.[25].
Before these materials can be introduced into aircraft structures, it is necessary to either develop a basic understanding of the influence of cavitation on mechanical properties, or develop a technique for eliminating it. One method for eliminating cavities involves the incorporation of a superimposed hydrostatic back pressure during the forming cycle [10,11,26,27] which suppresses cavity formation and improves the alloys formability (Figure 9b).

3.1 (contd.......)

The following sections address the effect of superplastic strain on the mechanical properties of a series of formed alloys, with and without cavitation being present.

3.2 Tensile Property Data

The effect of superplastic strain and test temperature on the tensile properties of a range of Supral alloys which have been formed without back pressure is presented in Table 3. It is apparent that the alloys are reasonably isotropic and that strength and ductility are reduced with increasing strain, which results in increasing cavitation. The clad alloys are also weaker than the unclad version. This is due to two factors: the presence of a weaker commercially pure aluminium layer and the increased levels of cavitation produced in these alloys with superplastic strain [17, 18].

Cavitation created during the forming cycle may be sealed by a post form hipping [28,29] operation, but any improvement in properties appears to be limited to removal of small cavities, rather than the large interlinked cavities seen at high strains. Moreover, cavity closure does not necessarily indicate that the cavities have been sealed. In Figure 10, post form hipping of Supral 150 achieved parent densities over a range of hipping temperatures [29], but some cavitation returned after re-solution treatment (as indicated by a drop in density) even when hipping temperatures of 530°C were used. It is thought that this is caused by the presence of hydrogen in the original cavities, or its subsequent migration to them. Vacuum degassing prior to hipping does reduce the re-appearance of cavities and enables parent densities to be achieved after resolution treatment. Sealing of cavities in this manner increases the ductility of the formed material, but very high hipping pressures (100MPa) and long times (3 hours) are required before improvements in the tensile strength are seen.[19].

Further improvements in post-formed properties can be achieved by suppressing cavity formatior during forming and typical properties for a range of alloys are presented in Figure 11. Test specimens have been extracted from test boxes biaxially formed with the benefit of back-pressure. As a rule, tensile strength reduces with increasing superplastic strain. The initial rise in strength for the Supral 220 material is due to the onset of dynamic recrystallisation during the initial stages, refining grain size to the order of 4-6μm. This can clearly be seen by a comparison of the microstructure for undeformed and deformed material in Figure. 9c. [21].

It is interesting to note that, with the exception of Al-Li alloy (Alcan 8090), it is necessary to resolution treat prior to ageing in order to obtain maximum strength. A lack of quench rate sensitivity is extremely important. It is possible to achieve near maximum properties for the 8090 alloy without the problems of distortion that are associated with re-solution treatment and quenching of thin complex shaped components. The 8090 material also offers 10% reduction in density over the other alloys and an elastic modulus of 78GPa. A high modulus is particularly important for thin sheet structures which are designed on stiffness rather than strength.

3.3 Fatigue

The effect of superplastic strain on the fatigue performance of Supral, 7475 and Alcan 8090 alloys has been determined by a number of people [11,13,14,15,17,18,22]. All the data has been generated using sinusoidal loading and a stress ratio (R) of 0.1. The fatigue performance of Supral (Figure 12) and 7475 alloys (Figure 13) has been found to be independent of test orientation, but longitudinal data is only available for the 8090 alloy (Figure 14).

The two sets of results for the Supral alloys in the T6 condition (Figure 12) show marked differences in their fatigue performance even though the material tested was supplied and formed by the same manufacturer. The data reported by Shakesheff [17,18], gives a consistently better fatigue performance for the four alloys, which appears to be less dependent upon su erplastic strain than that reported by Lipsius et al [22]. In Supral 100 and 150 this difference could be associated with different levels of intergranular cavitation, which act as fatigue initiation sites. Unfortunately, this cannot be confirmed as no cavitation data was reported by Lipsius et al. This was not the case for the clad Supral 220 alloy, as both workers reported similar levels of cavitation (6-7% after 250% superplastic strain). A more likely explanation in this instance is that it is due to incipient melting. Shakesheff established that incipient melting occurred at 527°C [18], which is 3°C lower than the recommended solution heat treatment temperature of 530°C. As a consequence Shakesheff reduced the solution heat treatment temperature to 520°C, and did not observe fatigue initiation from sub-surface defects as reported by Lipsius et al.

The effect of superplastic strain on notched (Kt=3) fatigue has only been determined for Supral 220. The data reveals a slight reduction in the overall fatigue performance in comparison to unnotched data, with a lower degree of scatter and an independence of superplastic strain up to at least 250%. The reduced scatter is consistent with the notch stress intensity overriding fatigue initiation from the clad surface, or sub-surface defects associated with liquation and/or cavitation.

3.3. contd.......

Comparison of the results for the low strength (Supral 100 and 150) and high strength (Supral 220) alloys clearly demonstrates that the clad version has poorer fatigue performance. This is attributed to fatigue initiation in the clad layer, especially as the clad grain size increases more rapidly than the matrix during superplastic deformation. It would also explain why the fatigue performance of clad Supral 220 is apparently insensitive to cavitation whilst the unclad version is cavitation dependent.

Information in the literature on the unnotched fatigue performance of 7475 in the T6 condition is conflicting. In Figure 13a, Bampton et al [15] report that the fatigue performance is improved with superplastic strain, although actual strain levels are not quoted. On the other hand, Bampton and Edington [13] (Figure 13b) reported that the fatigue life for a peak stress level of 275MPa and R of 0.1 decreased with superplastic strain in excess of 100%. No explanation is available. When formed under a back pressure the fatigue of material strained up to at least 150% was the same as that for undeformed fine grained sheet [14].

The fatigue performance of the 8090 alloy after superplastic forming using back pressure and immediate post form ageing (no resolution treatment) to a T6 state is given in Figure 14.[4] The reduction in fatigue performance at strains in excess of 100% is due to the original unrecrystallised microstructure gradually being converted to a weaker recrystallised microstructure. However, limited data for material which has been re-solution treated prior to ageing indicates that fatigue properties can be improved to give results which are comparable to undeformed, heat cycled and aged material as well as 7475-T6 and Supral 220.

3.4 Crack Propagation/Fracture Toughness

Data on the fatigue crack growth rate of the Supral alloys in the as-formed and formed and aged condition, is presented in Figures 15 and 16.[17,18] No back-pressure was used. Generally, the unclad Supral alloys have a lower fatigue crack growth rate at the same ΔK and superplastic strain, whilst the fatigue crack growth rate increases with increasing superplastic strain. Post forming heat treatment reduces the threshold ΔK from 4.5-6.5 MPa $\sqrt{m}$ for the as formed material (Fig.15), to 3.0-3.8 MPa $\sqrt{m}$ for the formed and heat treated condition (Fig.16). Heat treatment also increases the fatigue crack growth rate for material which has been subjected to 250% strain : the 100% strained material is comparable to the as formed condition.

It is not known whether the dependence of fatigue crack growth rate on superplastic strain would still exist if back pressure was used, although limited data for cavitation free 7475 material subjected to up to 150% strain and heat treated to T6 condition [14] does suggest that fatigue crack growth rate is independent of superplastic strain. Agrawal and Truss [14] also noted that the fatigue crack growth rate in 7475 is increased by 8-10% when the superplastically formed material was tested under the same conditions but in salt water rather than air.

The only fracture toughness data that is available is for clad and unclad Supral 220 material [19]. This is summarised in Figure 17, and shows that Kc reduces with increasing superplastic strain and that the clad sheet has a lower Kc than the unclad sheet for a given strain, probably due to a higher degree of cavitation. The importance of cavitation is confirmed by the hipped results, where an approximate increase of 25% on the Kc value for unclad Supral 220 was achieved by hipping prior to heat treatment in order to remove the 3-4% cavitation that was present.

3.5 Compressive and Bearing Strength

The compressive strength of the Supral alloys [22] is independent of superplastic strain up to approximately 250% whereas the 7475 [14] alloy begins to fall after approximately 100%. On the other hand, the bearing strength of Supral [22] and 7475 [14] alloys were generally reduced with increasing superplastic strain, although some anomalies were observed with the Supral alloys.

CONCLUSION

The mechanical properties of both titanium and aluminium alloys tend to be reduced by superplastic strain. In titanium alloys, the reduction in strength is primarily due to variation in microstructure resulting from the thermal cycle and to a lesser extent by the superplastic strain. However, increasing the amount of superplastic deformation will make a material with an aligned microstructure or strong crystallographic texture more isotropic.

The aluminium alloys are susceptible to intergranular cavitation during superplastic deformation, and the amount of cavitation increases with increasing strain. The presence of these cavities results in a significant reduction in mechanical properties and optimum design parameters can only be achieved if cavity formation is suppressed during the deformation cycle.

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N.B. An addendum follows at page 6-24.

HEAT TREATMENT	Orientation	0.1PS MPa	0.2PS MPa	0.5PS MPa	TS MPa	E GPa	e_u on 20 mm %	e_t on 25 mm %
Mill annealed (as received)	L	989	997	1011	1079	113	4.2	13.5
	T	1034	1052	1060	1060	123	5.8	14.3
900 °C ½ hr cooled 25°C/min	L	953	948	948	1014	107	5.8	18.8
	T	1025	1033	1034	1034	112	5.6	11.9
900°C ½ hr cooled 25°C/min ,500°C 24 hrs	L	1033	1028	1033	1118	111	4.9	14.5
	T	1096	1104	1111	1121	121	5.6	12.4
900°C ½ hr cooled°C/min	L	997	1013	1027	1116	108	5.0	15.0
	T	1062	1088	1109	1125	118	4.3	13.2
900°C ½ hr cooled 150°C/min ,500°C 24 hrs	L	1070	1072	1082	1175	114	4.9	13.6
	T	1129	1142	1155	1174	122	4.8	12.5
900°C 3½ hr cooled 25°C/min	L	922	922	928	1017	108	6.2	18.1
	T	1002	1016	1025	1044	121	6.0	17.8
900°C 3½ hr cooled 150°C/min	L	963	984	1000	1096	110	5.3	18.1
	T	1023	1055	1082	1117	119	5.4	15.9

PS Proof Stresses, 0·1%etc., ε_u Uniform, ε_t Total Elongations

Each result is the average of two tests

ROOM TEMPERATURE TENSILE PROPERTIES OF IMI 550 SHEET — TABLE 1

POST – FORMING HEAT TREATMENT	Orientation	SP Strain %	0.1PS MPa	0.2PS MPa	0.5PS MPa	TS MPa	E GPa	e_u on 10 mm %	e_t on 12 mm %
Cooled 25°C/min	L	91	899	907	917	994	105	3.9	7.6
		123	931	935	941	1027	104	5.2	8.7
		190	832	841	848	926	98	5.3	7.5
	T	37	961	959	959	1020	112	5.2	8.8
		82	941	947	959	1001	111	3.5	4.8
		120	926	921	922	978	105	4.6	7.8
		169	960	965	972	1003	121	2.7	5.9
		172	963	974	963	1019	113	6.0	7.5
Cooled 25°C/min 500 °C 24 hrs *	L	172	963	970	979	1058	108	5.5	10.4
Cooled 150°C/min	L	181	907	923	957	1035	98	3.0	10.8
Cooled 150°C/min 500 °C 24 hrs	L	183	1004	1020	1040	1127	108	5.4	9.5
900°C 24 hrs cooled°C ½ hr, 500 °C 24 hrs	L	191	997	1017	1037	1123	112	3.0	9.0

* – Average of two results

EFFECT OF SUPERPLASTIC STRAIN AT 900 DEG. C AND POST-FORMING HEAT TREATMENT ON THE ROOM TEMPERATURE TENSILE PROPERTIES OF IMI 550 SHEET — TABLE 2

SPF STRAIN	TEST TEMPERATURE °C								
	-54			RT			90		
	0.2% PS MPa	TS MPa	% EL	0.2PS MPa	TS MPa	% EL	0.2PS MPa	TS MPa	% EL
	SUPRAL 100 (UNCLAD)								
	LONGITUDINAL								
L	318	426	12.5	304	409	12.5	284	368	15.5
M	303	405	11.5	304	407	10.7	289	360	13.0
H	258	356	7.8	292	392	7.7	247	310	6.3
	TRANSVERSE								
L	314	429	10.8	311	423	11.7	302	391	13.0
M	326	424	7.3	322	436	9.3	298	382	12.8
H	346	431	4.3	302	391	5.1	250	303	4.0
	SUPRAL 150 (CLAD)								
	LONGITUDINAL								
L	275	396	12.3	293	397	11.0	277	365	10.4
M	300	394	10.2	272	364	8.5	270	362	9.7
H	247	293	3.0	254	316	5.3	245	300	2.8
	TRANSVERSE								
L	306	392	8.6	296	393	8.2	325	375	9.3
M	283	369	7.1	285	375	8.0	272	358	7.9
H	261	311	2.3	263	307	2.7	242	273	1.7
	SUPRAL 220 (CLAD)								
	LONGITUDINAL								
L	362	427	7.1	351	414	8.1	333	375	10.5
M	379	447	7.3	345	404	7.1	337	367	7.4
H	327	367	3.0	331	356	2.3	298	317	2.7
	TRANSVERSE								
L	349	394	4.4	363	423	7.4	328	378	8.2
M	369	424	5.4	350	412	7.5	330	378	6.7
H	—	—	—	327	347	3.0	—	—	—

L = 25 – 75% M = 75 – 150% H = 150 – 250%

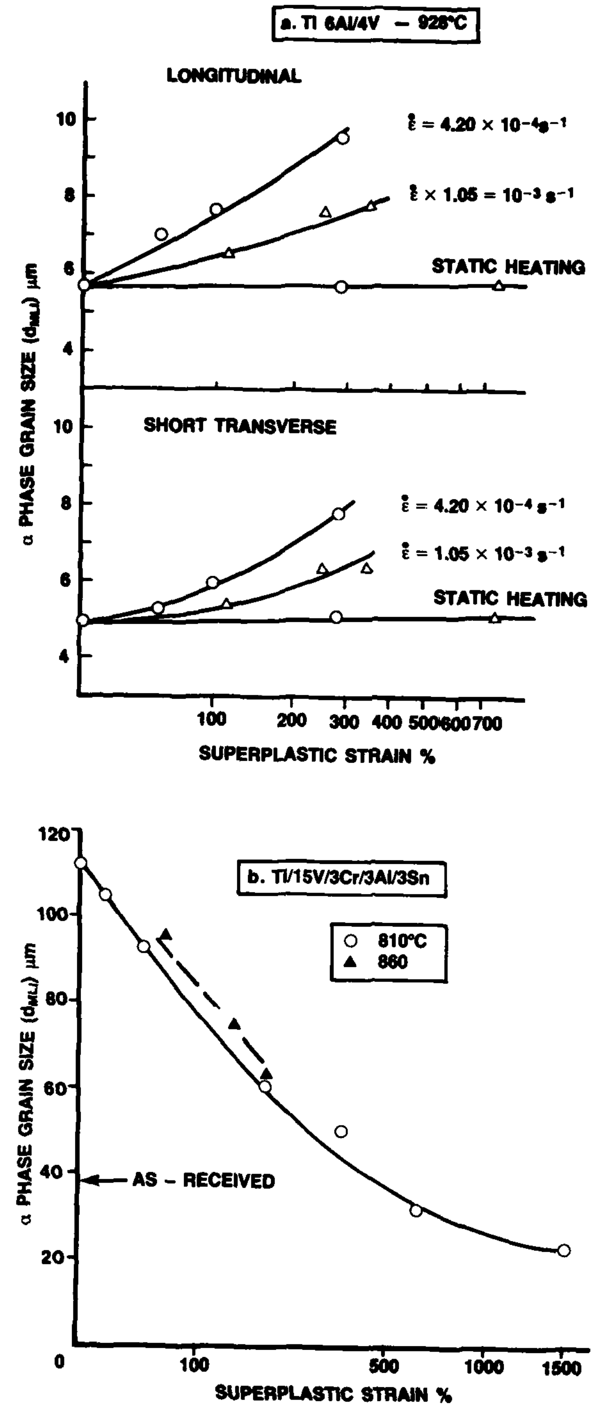
COMPOSITION (WEIGHT %)

ALLOY	Cu	Mg	Zr	Fe	Zn	Ge	Si	Mn	Ni	Cr	Al
Supral 100	6.0	0.25	0.4	0.1	0.05	—	—	< 0.1	—	—	Remainder
Supral 220	5.9	0.36	0.4	0.18	0.07	0.1	0.12	< 0.01	< 0.01	< 0.02	Remainder

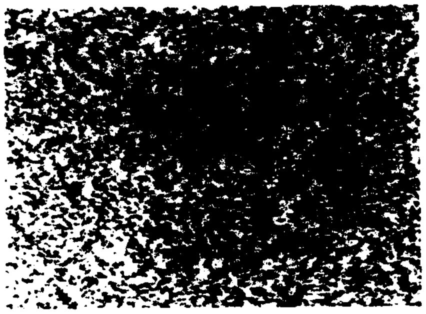
SUPRAL 150 IS A CLAD VERSION OF SUPRAL 100

EFFECT OF TEST TEMPERATURE AND SUPERPLASTIC STRAIN WITHOUT BACK PRESSURE ON TENSILE PROPERTIES OF SUPRAL ALLOYS —T6 CONDITION (Ref 22).

TABLE 3

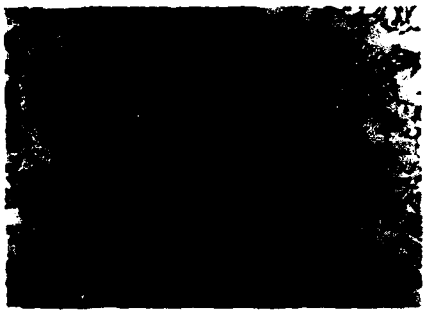


EFFECT OF STRAIN RATE AND STRAIN ON THE α GRAIN SIZE (d_{MLI})
IN THE L-ST PLANE AFTER DEFORMATION — Ti 6Al/4V & Ti/15V/3Cr/3Al/3Sn MATERIAL FIG.1

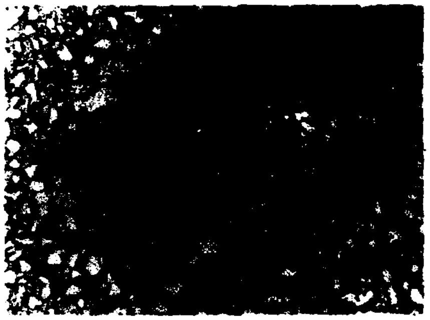


a) As Received (Acceptable)

b) As Received (Banded)



c) SPF with Mufflebox Technique

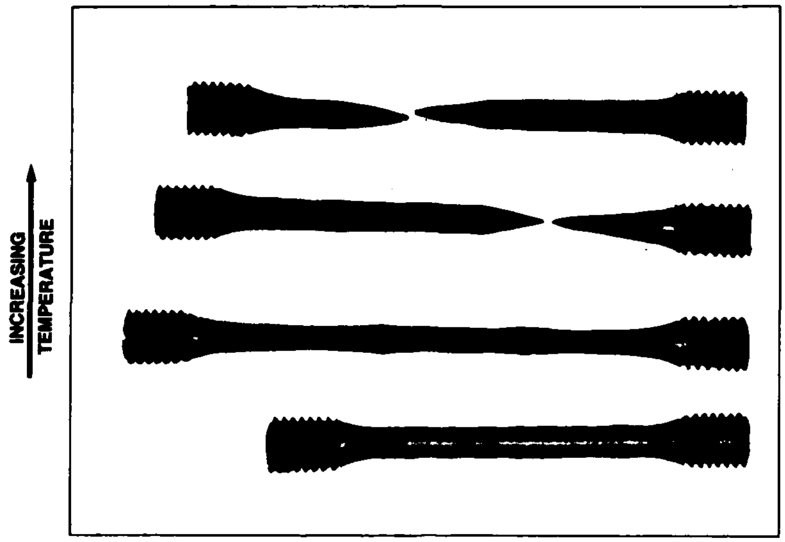


d) After SPF with Hot Open Die Technique

0.1mm

TYPICAL PRE AND POST FORMING MICROSTRUCTURES FOR Ti/6Al/4V MATERIAL

FIG.2

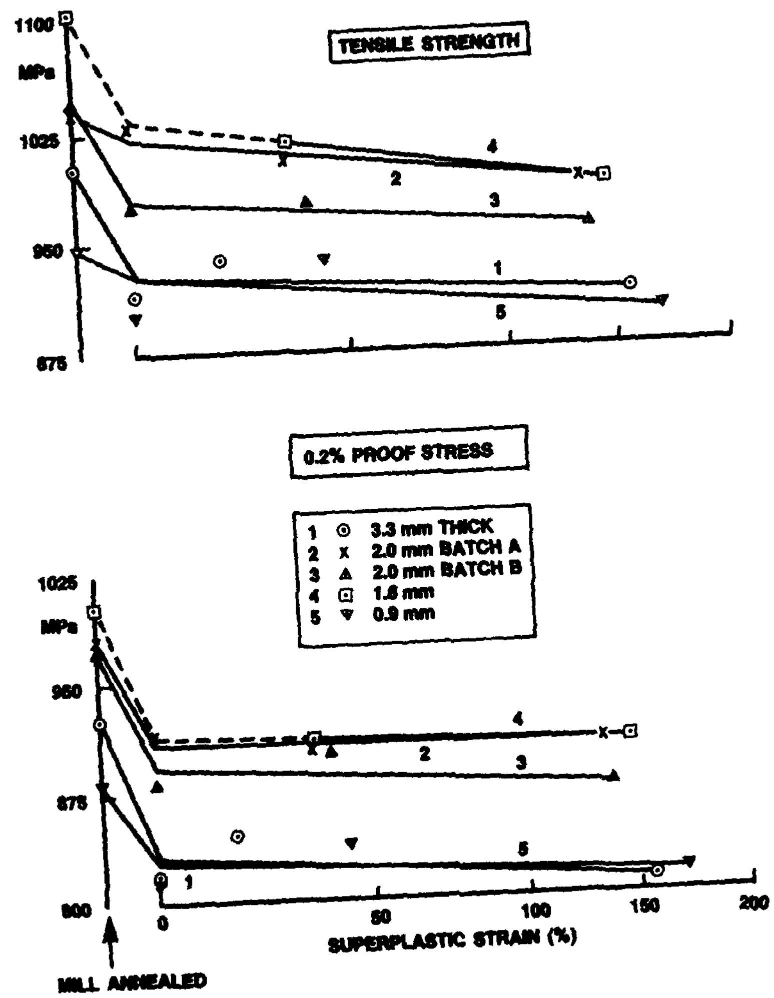


a) BAR MATERIAL

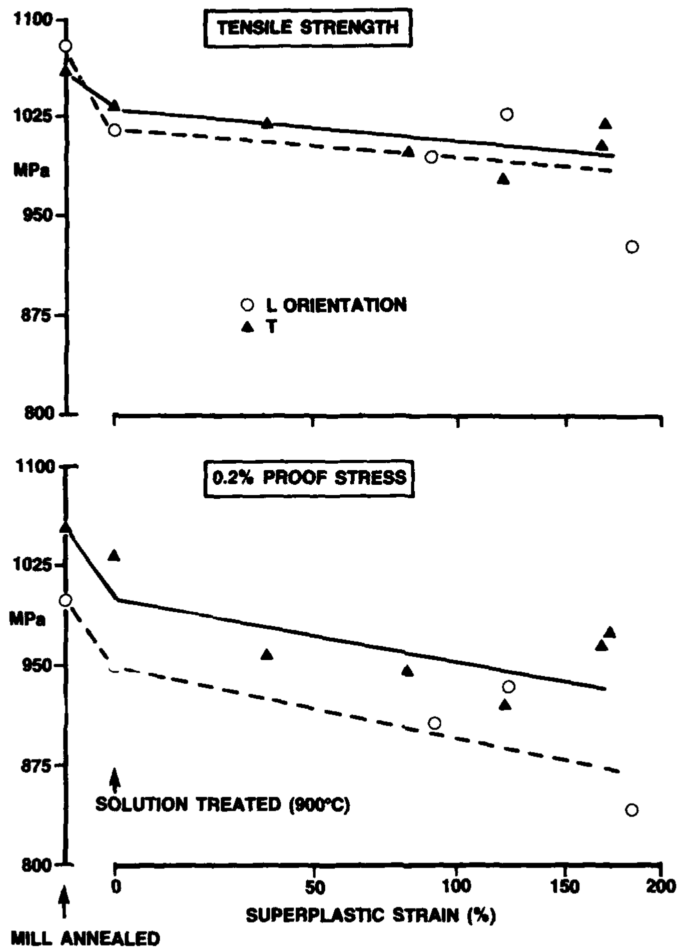
b) SURFACE OF SHEET MANUFACTURED FROM BAR MATERIAL

EFFECT OF ALIGNED ALPHA BANDS ON THE SHAPE CHANGES
IN Ti/6Al/4V MATERIAL WITH TEST TEMPERATURE

FIG.3

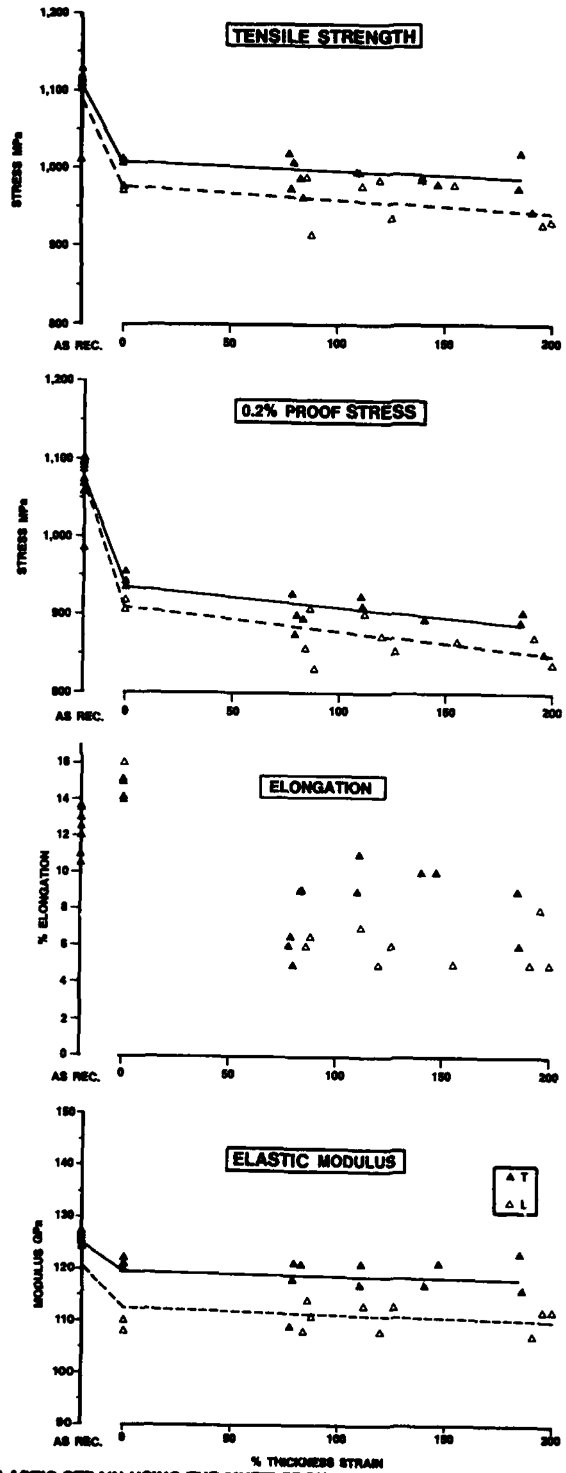


EFFECT OF SUPERPLASTIC STRAIN AT 925°C AND $3 \times 10^{-4}\ s^{-1}$ ON ROOM TEMPERATURE STRENGTH OF Ti-6Al-4V SHEET (L ORIENTATION)



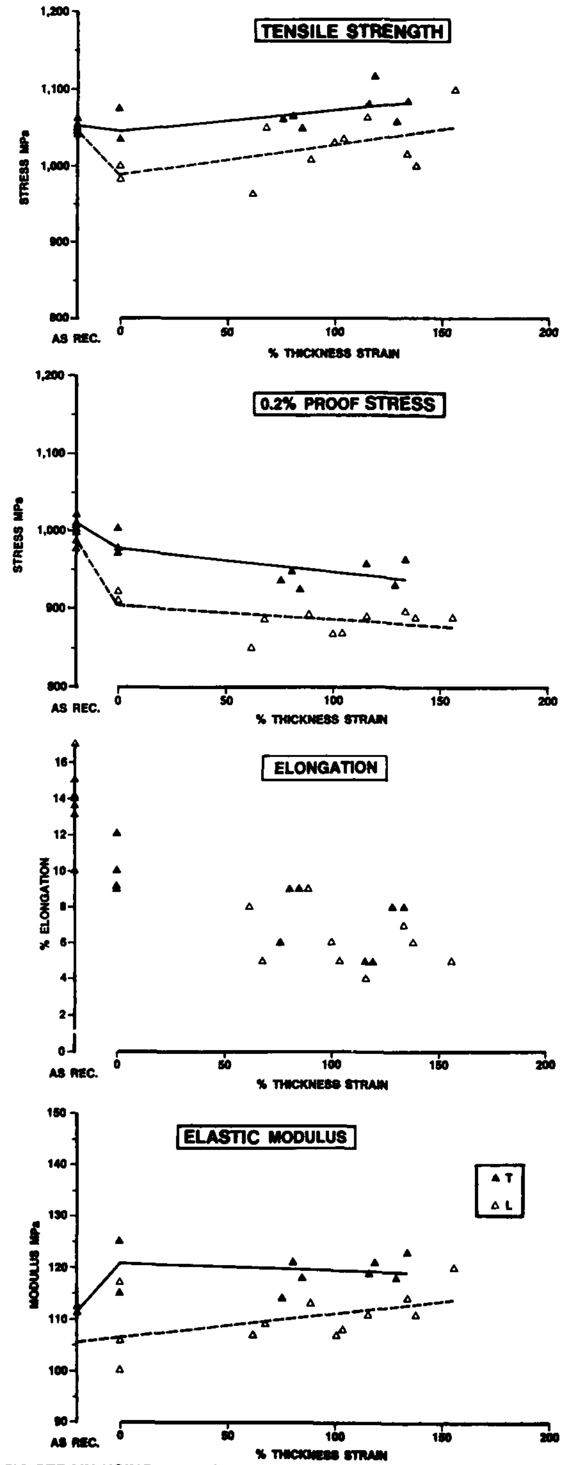
EFFECT OF SUPERPLASTIC STRAIN AT 900°C AND $3 \times 10^{-4}\ s^{-1}$ ON ROOM TEMPERATURE 0.2% PROOF STRESS AND TENSILE STRENGTH OF IMI 550 SHEET

FIG. 5



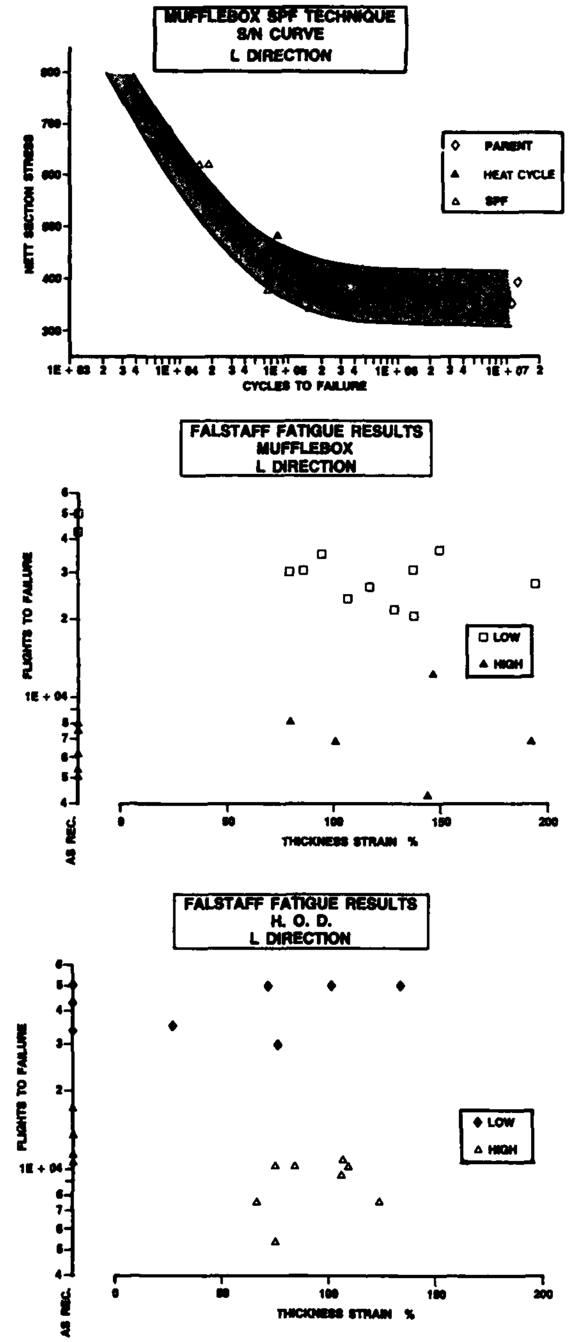
EFFECT OF SUPERPLASTIC STRAIN USING THE MUFFLEBOX TECHNIQUE ON TENSILE PROPERTIES (TEMP = 925°C, $\dot{\varepsilon} = 1.5 \times 10^{-4}$ sec^{-1})

FIG.6



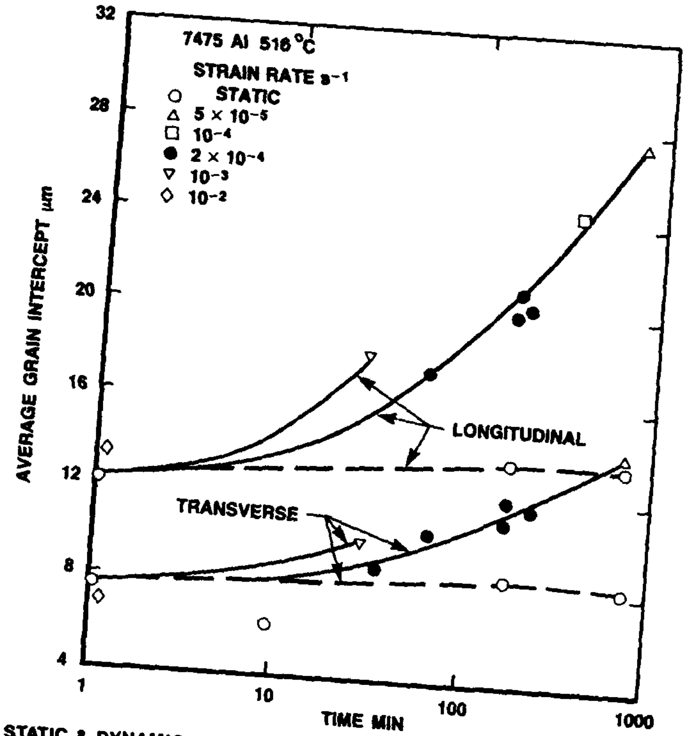
EFFECT OF SUPERPLASTIC STRAIN USING THE HOT OPEN DIE TECHNIQUE ON TENSILE PROPERTIES (TEMP = 925°C, $\dot{\varepsilon} = 1.5 \times 10^{-4}$ sec^{-1})

FIG.7



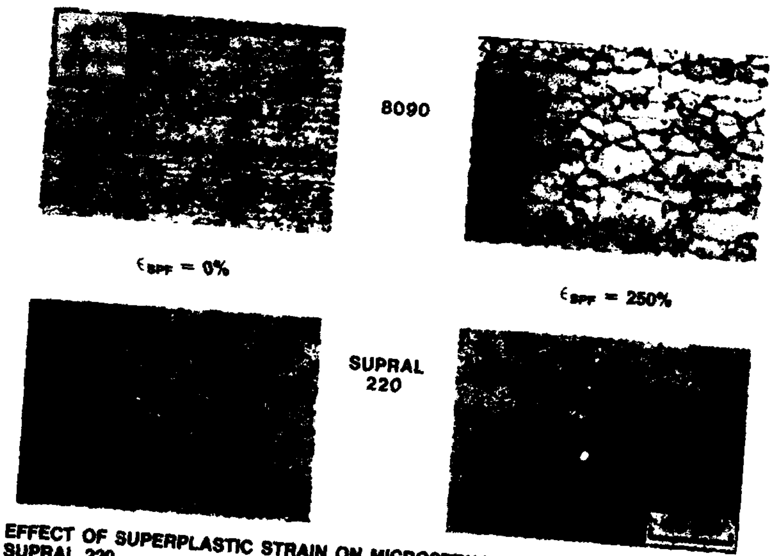
FATIGUE RESULTS FOR BIAXIALLY DEFORMED MATERIAL
USING BOTH MUFFLEBOX AND HOT OPEN DIE (HOD) TECHNIQUES

FIG.8



a) STATIC & DYNAMIC GRAIN GROWTH IN 7475 AL (Ref 16)

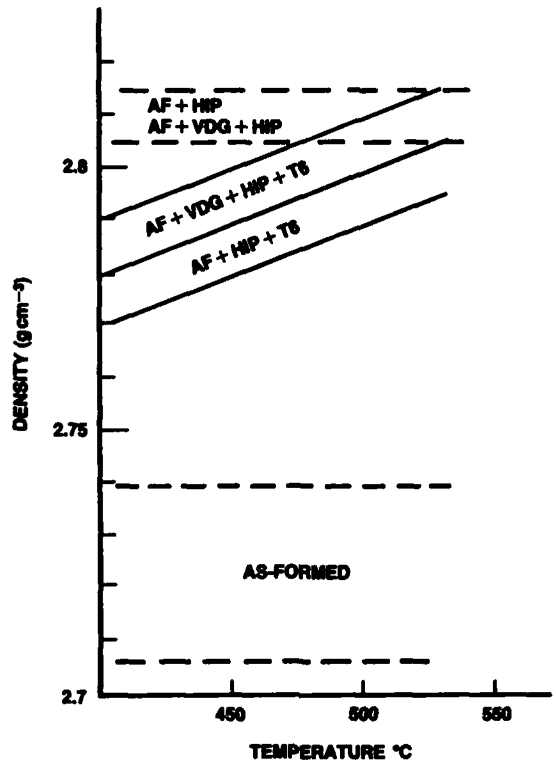
b) EFFECT OF BACK PRESSURE (MPa) ON CAVITATION IN SUPRAL 220



c) EFFECT OF SUPERPLASTIC STRAIN ON MICROSTRUCTURE IN 8090 & SUPRAL 220

THE EFFECT OF SUPERPLASTIC STRAIN ON THE MICROSTRUCTURE OF ALUMINIUM ALLOYS

FIG. 9



CONDITION		LONGITUDINAL 0.2% PS MPa	LONGITUDINAL TS MPa	LONGITUDINAL % EL	TRANSVERSE 0.2% PS MPa	TRANSVERSE TS MPa	TRANSVERSE % EL
AF		100	180	9	103	172	10
AF + T6		300	350	9	285	326	9
AF + HIP + T6	450C	234	336	12	231	332	12
	500C	238	336	14	233	330	14
	530C	235	317	10.5	222	315	10
	* 505C	294	391	12	305	410	14
AF + V + HIP + T6	450C	280	369	13	280	354	12
	500C	232	322	16	255	344	14

AF = AS FORMED

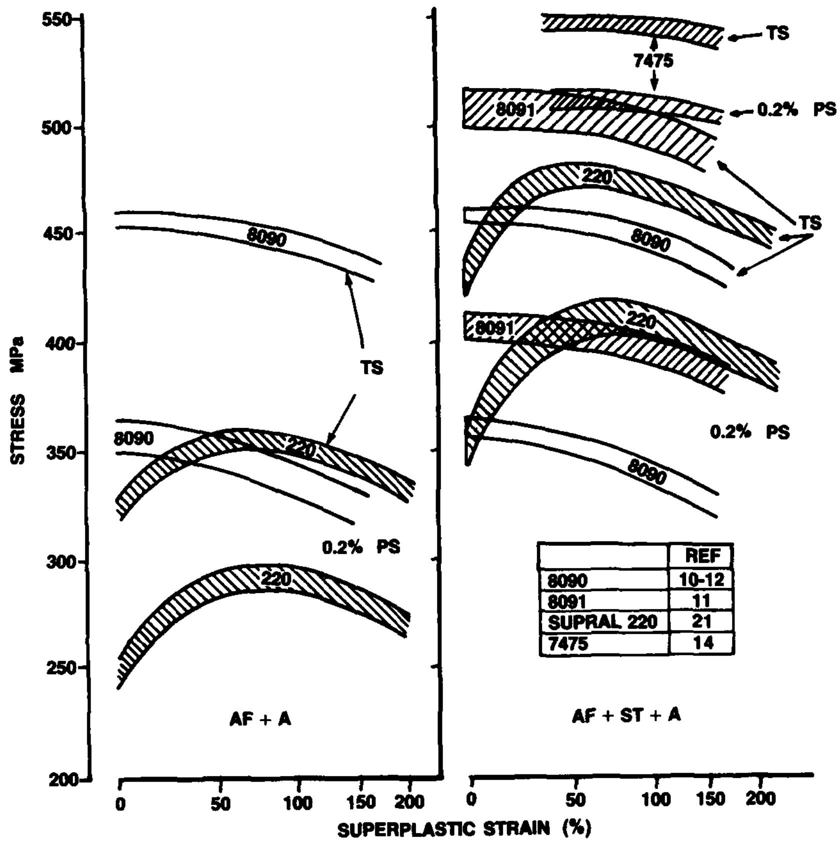
HIP = HIPPING FOR ½ at T°C AND 28 MPa PRESSURE

* 3 HRS AT T°C AND 100 MPa

VDG = VACUUM DEG AS FOR 1 HR AT 530°C AND 10^{-6} TORR

EFFECT OF POST FORM HIPPING & HEAT TREATMENT ON DENSITY/CAVITATION & TENSILE PROPERTIES OF SUPRAL 150 (CONTAINING 3% CAVITATION)

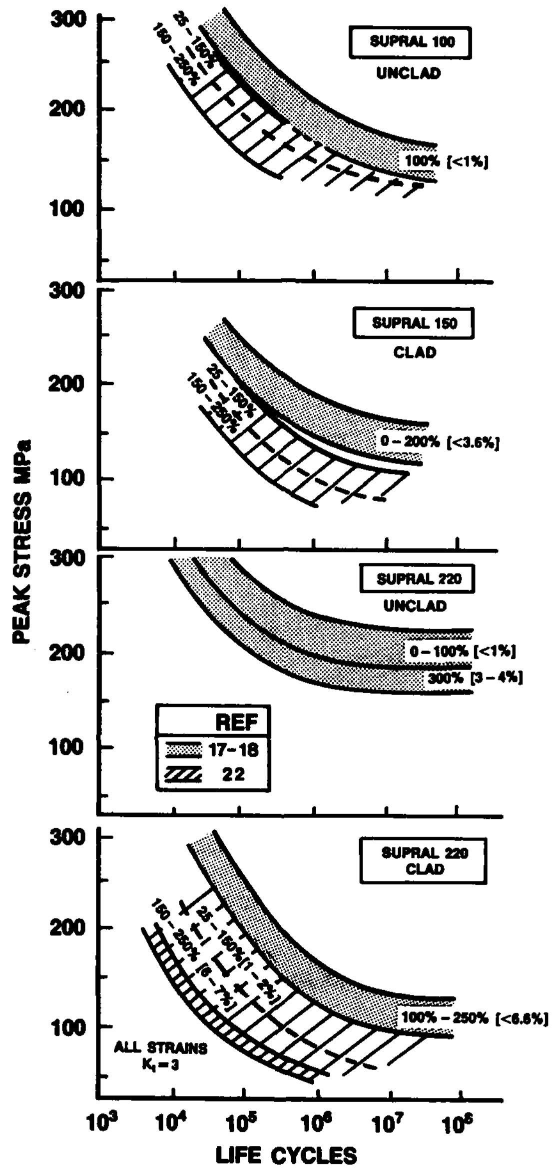
FIG. 10



	REF
8090	10-12
8091	11
SUPRAL 220	21
7475	14

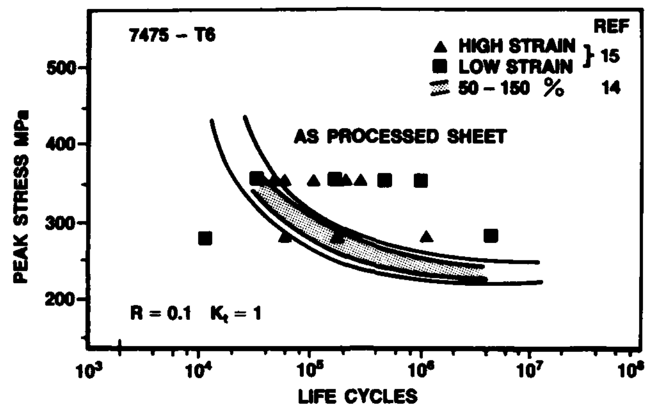
A = PEAK AGE TEMPER, AF = AS FORMED, ST = SOLUTION HEAT TREATMENT

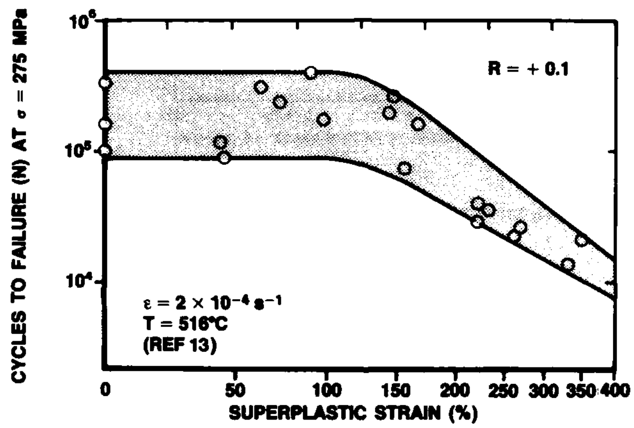
EFFECT OF SUPERPLASTIC STRAIN AND POST FORMING HEAT TREATMENT ON THE TENSILE PROPERTIES OF SUPERPLASTIC ALUMINIUM ALLOYS. **FIG. 11**



EFFECT OF SUPERPLASTIC STRAIN ON THE FATIGUE AT R = 0.1
OF SUPRAL ALLOYS FORMED WITHOUT BACK PRESSURE
AND POST FORM HEAT TREATED TO T6 CONDITION
(NUMBERS IN BRACKETS INDICATE PERCENT CAVITATION)

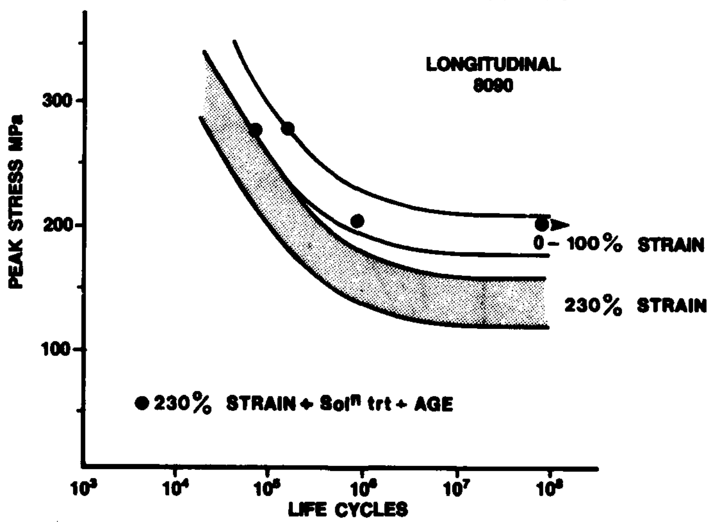
FIG. 12





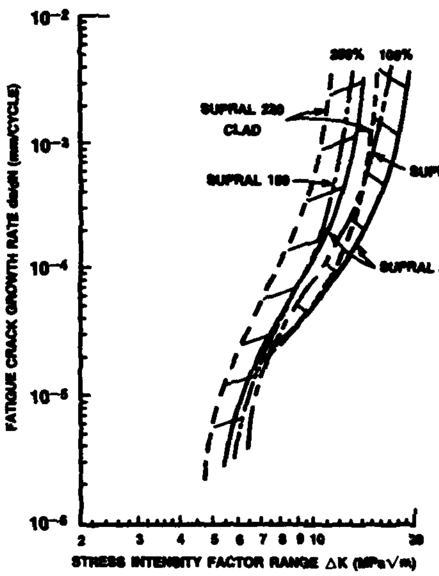
EFFECT OF SUPERPLASTIC STRAIN ON THE FATIGUE OF 7475 IN THE T6 CONDITION

FIG.13

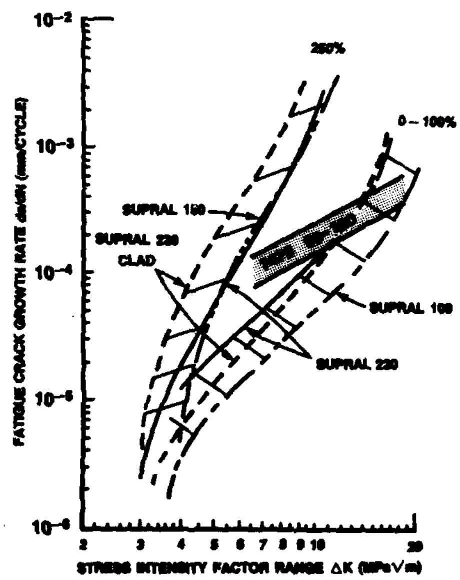


EFFECT OF SUPERPLASTIC THICKNESS STRAIN ON THE FATIGUE OF 8090 ALLOY BIAXIALLY FORMED UNDER A BACK PRESSURE OF 3·45 MPa AND AGED FOR 24H AT 185°C

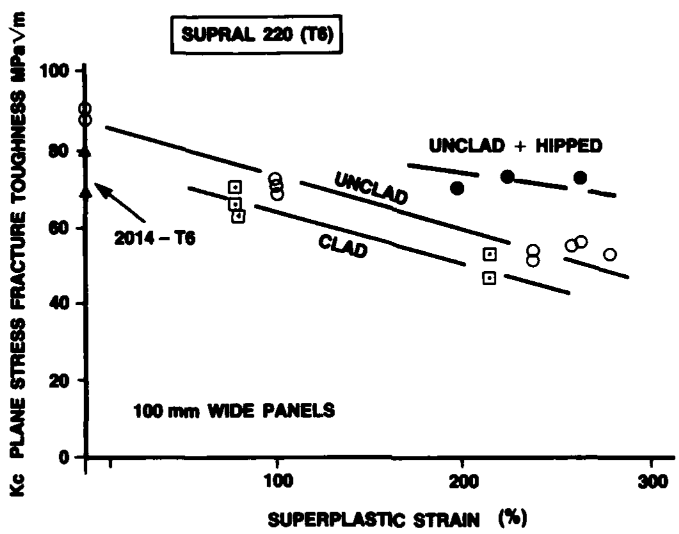
FIG.14



EFFECT OF SUPERPLASTIC STRAIN ON FATIGUE CRACK GROWTH RATES OF THE SUPRAL ALLOYS
(R = 0.1, Kt = 1) FIG.15



EFFECT OF SUPERPLASTIC STRAIN ON FATIGUE CRACK GROWTH RATES OF THE SUPRAL
ALLOYS AND 7475E AFTER SOLUTION HEAT TREATMENT AND AGEING (–T6) FIG.16



EFFECT OF SUPERPLASTIC STRAIN AND HIPPING FOR 90 MIN AT 360°C AND 103·5 MPa
ON THE L-T FRACTURE TOUGHNESS OF CLAD AND UNCLAD SUPRAL 220 SHEET IN -T6 CONDITION FIG.17

ADDENDUM TO 1987 AGARD LECTURE SERIES NO 154 ON MECHANICAL PROPERTIES OF SUPERPLASTICALLY FORMED TITANIUM AND ALUMINIUM ALLOYS

by

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A1 INTRODUCTION

This Addendum is intended to follow the 1987 Lecture on page 6.1 and to provide an update on developments since the last Lecture Series No 154 was published.

The three most important factors affecting post formed properties remain [1]:-

1 thermal cycles associated with superplastic forming (SPF),

2 plastic deformation,

3 reactions with the environment.

Further work has been reported on the effect of superplastic deformation on Ti-alloys in the form of sheet, bar and extrusions. Increasing interest in Al-Li alloys has led to more test data becoming available on the mechanical properties of these alloys after SPF. These topics are discussed in this Addendum.

A2 TITANIUM ALLOYS

A2.1 *Tensile Properties of Sheet, bar and extrusions*

The need to obtain a uniform distribution of equiaxed alpha and beta phases for isotropic superplasticity is now recognised [2]. The large variation in transverse strain produced in sheet by banded α or β microstructure is shown in Figure A1. This variation in strain can reduce the sheet thickness uniformity and adversely affect the mechanical properties.
Post formed sheet tensile properties have been summarised in the 1987 Lecture (p 6.2) and in subsequent papers [1, 2, 4]. The yield and tensile strength of post formed Ti-6Al-4V alloy sheet may be above or below the minimum specification (AMS4922) values. Ageing of alloys IMI 550 (Ti-4Al-4Mo-2Sn-0.5Si), Ti 6242 (Ti-6Al-2S-4Zr-2Mo) and Ti-6Al-4V after slow cooling from the forming temperature did not have any beneficial effect on sheet tensile properties. However complete re-solution heat treatment followed by ageing produced yield strengths comparable with those in the pre-formed state.

Data for Ti-6Al-4V alloy sheet in the thickness range 3.3 - 0.9mm were reported in the 1987 Lecture *(Figure 5). Additional data on the tensile properties of SPF Ti 6Al/4V in the form of* a histogram is given in figure A2 [3]. This data is taken from a range of SPF and SPF/DB production manufactured components, superplastic strains range from 0 to $\sim$ 150%. Further tests have been carried out on 0.5mm thick sheet after 300% extension to 0.24mm thickness [5]. The sheet deformed uniformly as shown in the section in Figure A3, but there was an increase in surface roughness as discussed elsewhere [6]. Tensile properties after deformation were 4-6% lower than those for as received sheet (Table A1) and the elongation was reduced by half, probably due to the reduced cross-section as found in biaxially formed sheet [7].

Rolled bar or extruded sections can also be superplastically formed with low flow stresses and high m-values (>0.6), but the large grain size or acicular microstructure in these product forms reduces the superplasticity compared with sheet [8]. For example, to obtain the same flow stress used for forming sheet the required strain rates ($\dot{\varepsilon}$) are in the ratio $\dot{\varepsilon}$ sheet: $\dot{\varepsilon}$ bar: $\dot{\varepsilon}$ extrusion $\simeq$ 2.4: 2.1: 1. A five-fold increase in forming time for extruded material may represent a significant increase in cost. Anisotropic superplasticity was observed in test pieces cut from an extruded section or cut from bar material in directions normal to the bar axis, but isotropic superplasticity was found in test pieces oriented parallel to the bar axis (see insets in Figure A4). This behaviour must be taken into account when forming shapes from these product forms. The microstructure of both bar and extrusion became more uniform and equiaxed after SPF (Figure A4). In the future the anisotropy effects may be reduced by modified processing and heat treatment of bar and extrusions prior to forming. The post formed tensile properties are shown in Table A1; a reduction of 2-7% was obtained in the bar properties but there was negligible change in the properties of the extruded material.

A detailed study of the superplastic deformation of electron beam welded Ti-6Al-4V sheet has been carried out [9]; the fusion (FZ) and heat affected zones (HAZ) were not superplastic. Nevertheless if the weld was aligned parallel to the principle strain axis the weld deformed with the sheet, weld undercuts were removed and relatively uniform sheet cross sections could be obtained if the weld was pre-machined (Figure 5). In the as welded state welds aligned parallel and in the centre of a tensile test piece gauge length caused high tensile strength. After SPF the strength was reduced by 8-15% (Table A1). It was concluded that after SPF the improved microstructure and the absence of residual stresses and weld undercuts would lead to weld fatigue properties close to those of the parent material.

A2.2 Elevated Temperature Properties

The elevated temperature properties of Ti-6Al-4V alloy sheet after SPF [10] was reduced in proportion to the room temperature strength reduction up to about 400°C (Figure A6). Tests in the temperature range 316-538°C have been reported for Ti-6242 sheet after SPF strains up to 700% [11]. The room temperature tensile and yield strengths were reduced by 10%. At 316°C creep tests at 427°C and 538°C showed the creep rate was reduced by superplastic strain, with the reduction greater after 200% strain than after 100% strain (Figure A7). The reduction was attributed to grain coarsening during the thermal cycle and for a given superplastic strain a greater creep resistance was expected for slower superplastic strain rates.

A2.3 Fatigue

Mutoh et al [12] have reported on the effect of 100% superplastic strain on the S/N fatigue (R=0.05) data for Ti-6Al-4V. Their results agree with previous work [13] that SPF or thermal cycling produces shorter fatigue lives and a reduction in the fatigue limit by 35-56% compared with the mill annealed material (Fig. A8).

However reheat treatment was found to increase the life (by a factor 1.4) to approach that of the as received material. Further work is required in this area.

A2.4 Fracture Toughness and Crack Propagation

Superplastic deformation appears to have no adverse effect on fracture toughness (FT) or crack propagation (CP) in Ti-6Al-4V [14]. For example FT values after SPF were in the range K_E = 1200 - 1500 $Nm^{-3/2}$ and after SPF + STA heat treatment K_E = 1400-1700 $Nmm^{-3/2}$, which is comparable to the as received values. CP values were also similar for forged and SPF materials [3].

A3. ALUMINIUM ALLOYS

In addition to data presented in Figure 11, 0.2% Proof Stress data on superplastically formed 2090 Aluminium/Lithium alloy in ST + A condition has been reported [15], this is slightly lower than that for 8090 (in the longitudinal test direction) data, e.g. 66% = 353 MPa, 111% = 330MPa, 233% = 325 MPa.

All 8090 data previously presented is for conventional coil rolled (CR) sheet, thermal cycled or superplastically formed under back pressure and subsequently aged for 24 hours at 185°C. However, the ageing treatment has now been changed to 32 hours at 170°C to be consistent with the current conventional 8090 sheet product heat treatment since there is no adverse effect on either the strength, or ductility.

Material processing route changes have also been instigated by Alcan International to improve the alloys SPF characteristics, designated here as 8090 (XR). Uniaxial and biaxial SPF trials conducted on 8090 (XR) sheet under back pressure [17] have demonstrated its improved formability, lower flow stress and improved as formed surface finish in comparison to that for 8090 (CR) sheet. The lower flow stress is significant in that a lower back pressure will be required to suppress the onset of cavitation.

A3.1 Tensile Properties

The 8090 (CR) sheet has pronounced tensile anisotropy which is not completely removed by SPF. The result of changing the processing route significantly reduces this anisotropy, as shown in Figure A9, [16]. The effect of quench rate on the tensile anisotropy for 8090 (XR) is presented in figure A10, [16, 17]. The low quench rate sensitivity of this alloy is a major benefit although it is dependent upon the alloy copper content, Figure A11, [18] and material thickness. The effect of superplastic strain on the tensile properties of 3mm and 6mm 8090 (XR) sheet is present in Table A2. For the 3mm sheet an increase in tensile properties, with no loss in ductility, is observed with increased superplastic strain whereas for the 6.0mm sheet there is a reduction in tensile properties. The loss in strength of the thicker gauge being indicative of it having a slower cooling rate after forming. As only a minimal drop in tensile properties occurs for sheet (<6mm) air cooled from the forming press this heat treatment practise is recommended. This heat treatment practise eliminates the need for re-solution treatment and possible distortion of thin complex shaped component on quenching.

A3.2 Fatigue Properties

A comparison between the fatigue performance of 8090 (CR) and 8090 (XR) material under flight by flight spectrum loading (FALSTAFF) [17] is presented in Table A3. The specimens had a stress concentration factor (Kt) of 2.5, and were tested with nett section stresses of 250MPa and 320 MPa at FALSTAFF load level 32. The results indicate improved fatigue lives for 8090 (XR) compared with 8090 (CR) material. The general reduction in fatigue properties observed with superplastic strain may be attributed to a reduction in static properties and/or a change in microstructure with superplastic strain.

A3.3 Crack Propagation

Data on the fatigue crack growth rate (FCGR) of 8090 (CR), thermal cycled and superplastically formed using 'back pressure', [1] is presented in figure A12. The FCGR for the SPF 8090 (CR) sheet is slightly lower than that for underformed material, and much lower than that previously reported for the Supral alloys (Figures 15 and 16), in the latter case however no 'back pressure' was used during the forming and cavitation was present.

A4. FUTURE MATERIALS

In recent years materials research and development, within the aerospace industry, in the area of metal matrix composites (MMC) has been steadily increasing. MMC's with high specific mechanical properties and improved elevated temperature performance offer further overall component weight reduction. Potential applications in sheet form would include aircraft skins, cockpit floors and general substructure. Of particular interest are particulate reinforced materials which can be combined with the SPF process to provide significant cost and weight savings in airframe component manufacture. Aluminium/Lithium 8090 particulate reinforced (17 v/o SiC) sheet material has been superplastically formed, under 'back pressure' [17] up to strains of 250% without microstructural cavitation. An aircraft demonstrator component manufactured from 8090 MMC is shown in Figure A13.

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MATERIAL	0.2% PS MPa	TENSILE STRENGTH MPa	YOUNG'S MODULUS GPa	TOTAL ELONGATION %
THIN SHEET (PARALLEL TO ROLLING DIRECTION) AS RECEIVED, 0.46 mm THICK	959	1004	115	11
AFTER SPF THERMAL CYCLE	896	994	123	12
AFTER 300% SPF EXTENSION AT 925°C AND $\dot{\varepsilon} = 6 \times 10^{-4}\ s^{-1}$ (0.24 mm THICK)	895	960	119	4.2
ROUND BAR (PARALLEL TO BAR AXIS) AS RECEIVED 4 mm DIA	916	1000	109	11
AFTER 300% SPF EXTENSION AT 925°C AND $\dot{\varepsilon} = 9 \times 10^{-5}\ s^{-1}$ (3 mm DIA)	895	930	121	5
α/β EXTRUSION (PARALLEL TO EXTRUSION DIRECTION) AS RECEIVED, 4 mm DIA	871	976	113	19
AFTER 300% SPF EXTENSION AT 925°C and $\dot{\varepsilon} = 7 \times 10^{-5}\ s^{-1}$ (2.5 mm DIA)	877	962	121	4.1
ELECTRON BEAM WELD IN 3 mm THICK SHEET (PARALLEL TO WELD DIRECTION) AS WELDED AND MACHINED (2.6 mm THICK)	906	1077	113	10.5
AFTER 300% SPF EXTENSION AT 925°C AND $\dot{\varepsilon} = 2 \times 10^{-4}\ 10^{-1}$ AND MACHINING TO 1 mm THICK	833	914	113	1.3

POST SPF TENSILE PROPERTIES OF Ti-6Al-4V IN VARIOUS PRODUCT FORMS TABLE A1

	TENSILE STRENGTH			0.2% PROOF STRESS		
CONDITION	L	60	LT	L	60	LT
3,0 mm						
AR-PQ-AGE	490	448	454	403	334	364
HC-AC-AGE	462	413	432	355	320	341
150%-AC-AGE	488	432	453	366	325	330
200%-AC-AGE	483		458	361		346
6,0 mm						
AR-PQ-AGE	464		463	376		356
HC-AC-AGE	436		437	343		320
150%-AC-AGE	419		415	316		309
200%-AC-AGE	424		420	310		307

NOTE: AGEING CARRIED OUT FOR 32 HOURS AT 170 DEG. C

MEAN STATIC TENSILE PROPERTIES FOR 8090 (XR) [17]

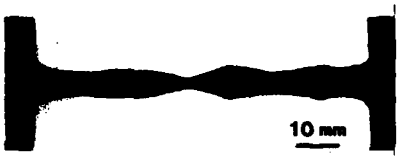
TABLE A2:

	8090 (CR)		8090 (XR)	
	250 MPa	320 MPa	250 MPa	320 MPa
3 mm MATERIAL				
AR-ST-PQ-AGE			21.9	9.1
HC-AC-AGE	27.0	7.4	31.9	8.4
100%-AC-AGE	14.6	6.6		
150%-AC-AGE				
200%-AC-AGE			20.0	3.4
6 mm MATERIAL				
AR-ST-PQ-AGE	-	-	>44.3	10.9
HC-AC-AGE	-	-	26.0	8.1
150%-AC-AGE	-	-	21.0	6.1
200%-AC-AGE	-	-	15.4	3.3

NB VALUES ARE GEOMETRIC MEAN OF A NUMBER

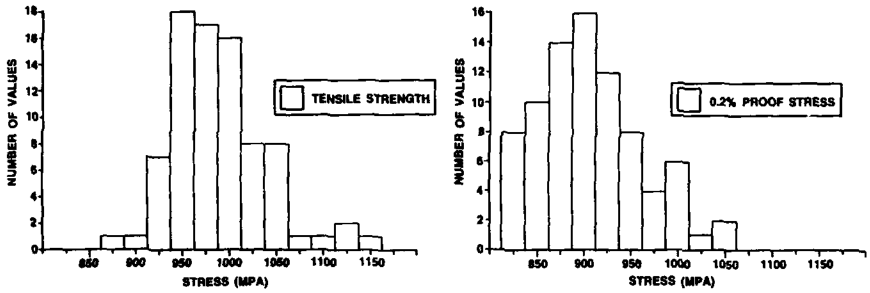
FATIGUE LIVES OF 8090 (CR) AND 8090 (XR) UNDER FLIGHT BY FLIGHT SPECTRUM LOADING (FALSTAFF) [17]. AGED 32h/170°C.

TABLE A3:



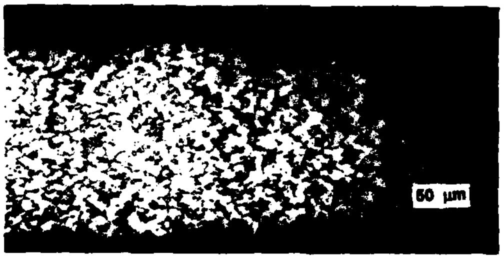
TI-6Al-4V SHEET TEST PIECE MACHINED FROM ROLLED BAR AND SUPERPLASTICALLY DEFORMED TO 273% ELONGATION AT 875°C. BONDED MICROSTRUCTURE AT A IS LESS SUPERPLASTIC.

FIGURE A1:



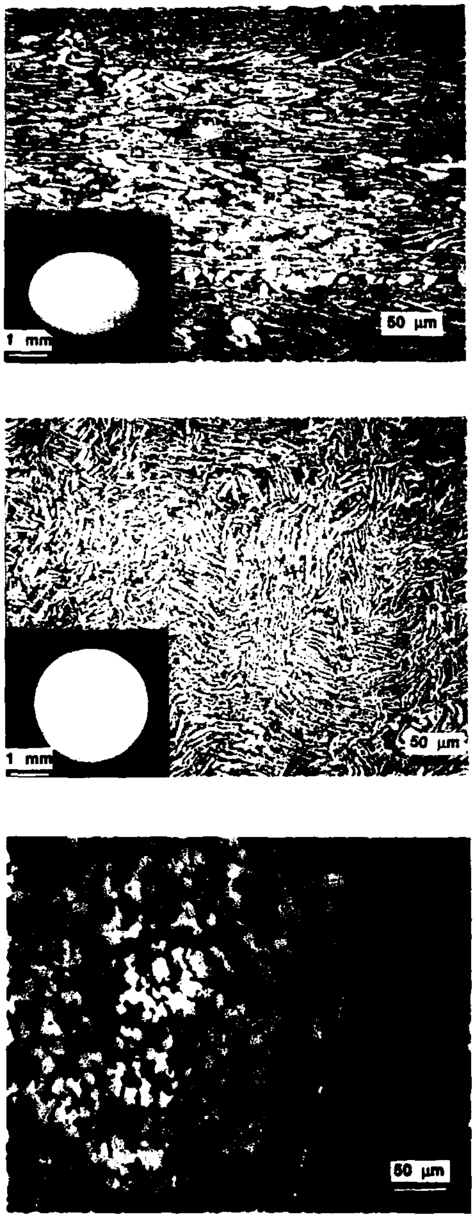
STATIC PROPERTY DISTRIBUTION SPF TA59 MATERIAL

FIGURE A2:



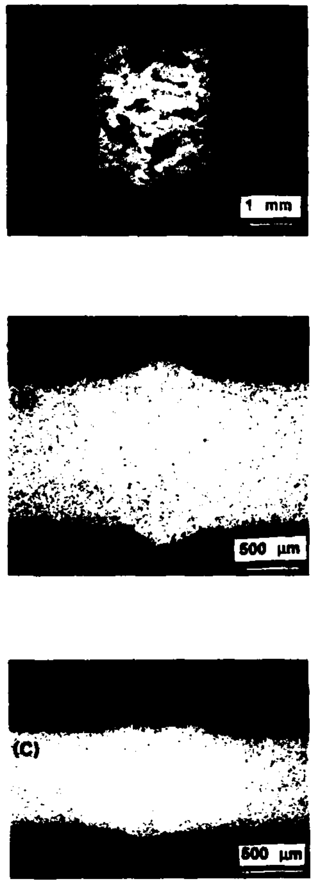
TI-AI-4V THIN SHEET (0.5 mm) AFTER 300% EXTENSION AT 925°C.

FIGURE A3:



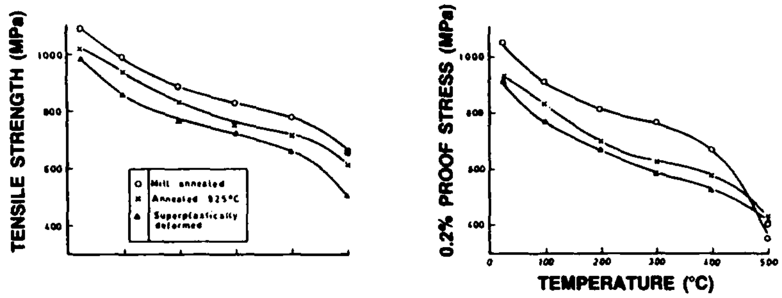
ROLLED Ti-6Al-4V BAR SHOWING INITIAL MICROSTRUCTURE (A) PARALLEL TO BAR AXIS (B) IN BAR CROSS-SECTION. INSETS ARE SHAPES OF BAR CROSS-SECTION AFTER (A) ε = 2.24 (B) 2.07 SUPERPLASTIC STRAIN. (C) MICROSTRUCTURE PARALLEL TO BAR AXIS AFTER ε = 2.24 AT 925°C.

FIGURE A4:

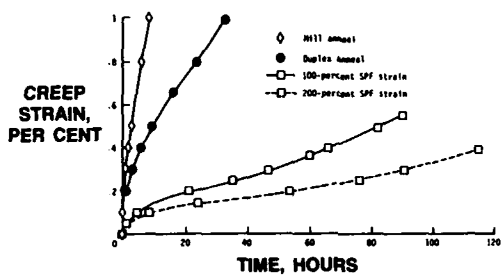


ELECTRON BEAM WELDED 3 mm THICK Ti-Al-4V SHEET CROSS-SECTIONS (A) AS WELDED (B) AFTER 300% SUPER-PLASTIC ELONGATION PARALLEL TO WELD AT 925°C BEFORE AND (C) AFTER REMOVING UNDERCUTS BY MACHINING.

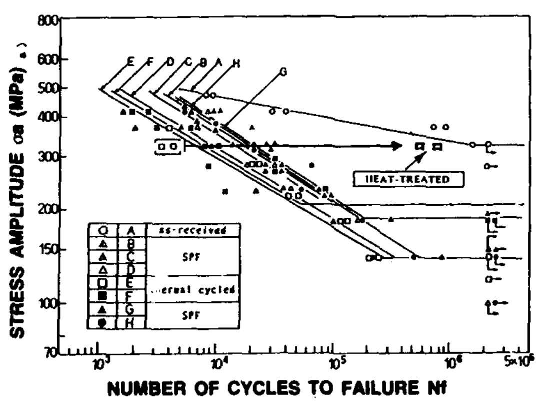
FIGURE A5:



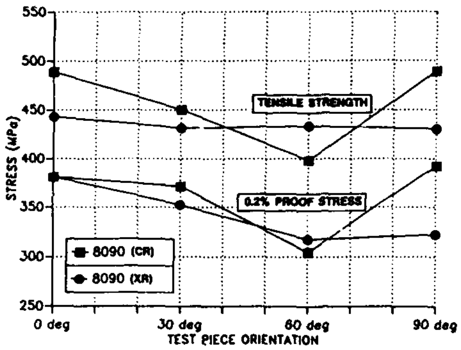
EFFECT OF SUPERPLASTIC STRAIN AT 925°C ON ELEVATED TEMPERATURE TENSILE AND 0.2% PROOF STRESS OF Ti-6Al-4V SHEET. **FIGURE A6:**



EFFECT OF SPF AT 900°C ON CREEP BEHAVIOUR OF Ti 6242 AT 537°C AND 276 MPa [10]. **FIGURE A7:**



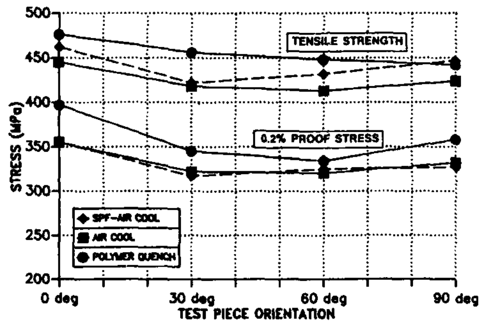
S/N CURVE FOR R = 0.05 AND 3 mm THICK Ti-6Al-4V SHEET AFTER 100% SUPERPLASTIC STRAIN AT 900°C. ARROW INDICATES EFFECT OF POST THERMAL CYCLE RE-HEAT TREATMENT [11]. **FIGURE A8:**



NB HEAT TREATMENT - SOLUTION TREAT/POLYMER QUENCH/AGE AT 170°C FOR 32 HOURS

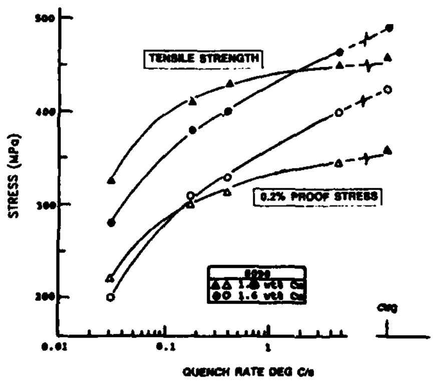
ROUND THE CLOCK TENSILE PROPERTIES FOR 8090 (CR) AND 8090 (XR) SHEET MATERIAL (1.6 mm) [16]

FIGURE A9:



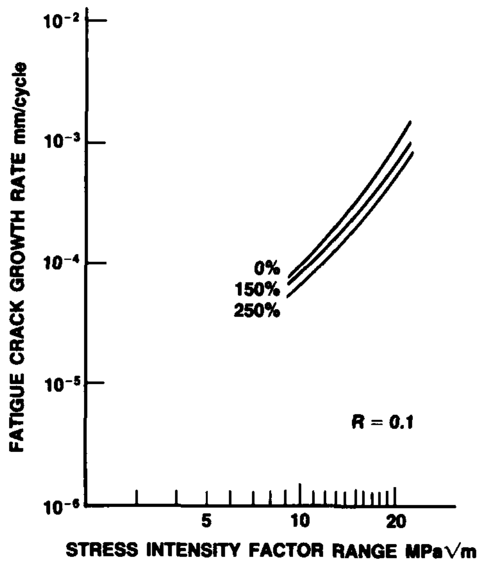
EFFECT OF COOLING RATE ON THE TENSILE ANISTROPY OF 3mm 8090 (XR) SHEET [16, 17]. AGED 32h/170°C

FIGURE A10:



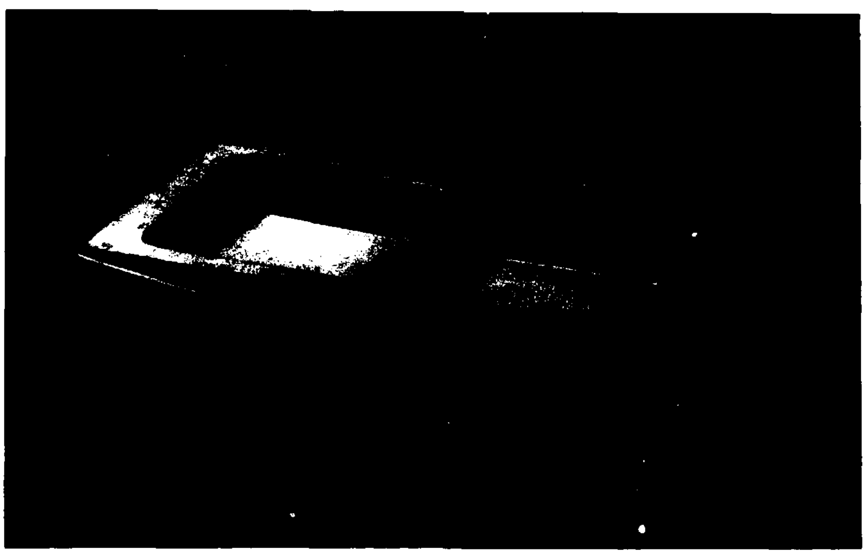
EFFECT OF QUENCH RATE AND Cu CONTENT ON STRENGTH OF 3 mm 8090 SHEET. (AGED 24h/185°C)

FIGURE A11:



EFFECT OF BIAXIAL SUPERPLASTIC STRAIN ON THE FATIGUE CRACK OF 3 mm 8090 (CR) FORMED USING BACK PRESSURE [1]. AGED 24h/185°C

FIGURE A12:



SUPERPLASTICALLY FORMED DEMONSTRATOR COMPONENT, 8090 SIC PARTICULATE (17%) SHEET (MATERIAL COURTESY OF BRITISH PETROLEUM, RESEARCH CENTRE, SPF AT BAe. MILITARY AIRCRAFT DIV)

FIGURE A13:

Designing For Superplastic Alloys

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SUMMARY

Twenty years of development have brought the processes of superplastic forming and diffusion bonding to a state of maturity. These processes provide the Project Designer with the opportunity to design components on new projects which are both cost and weight efficient. However, to achieve optimum performance, the designer needs to have an in-depth understanding of the freedoms and limitations provided by these processes. Substantial evidence exists to support the claim that titanium alloys, when processed by the SPF/DB route, can compete in weight and more particularly cost, with conventional aluminium fabrication. This is likely to be a major factor in the future exploitation of these processes and clearly requires a revision to the designers traditional views of the areas of application for titanium alloys. The more recent developments in the processing of high strength superplastic aluminium alloys will clearly add to the further use of the superplastic forming process on future aerospace products, but the development of a combined SPF/DB process for aluminium alloys, with the full range of capabilities provided currently by titanium alloys, remains to be established.

1.0 INTRODUCTION

My colleagues in this lecture series have dealt with the metallurgical and property aspects of superplasticity and diffusion bonding and clearly these are key aspects in the design of superplastically formed (SPF) and diffusion bonded (DB) components.

This paper expands the design theme by reviewing the current status of the processes of SPF and DB with respect to their maturity as design/production processes and attempts to identify the specific design flexibilities and limitations which these processes provide.

For ultimate success, these processes need to achieve a status of being accepted by the Project Designer as providing efficient and reliable production components from both a structural and cost point of view and which can be produced in a timescale which is compatible with that for the project as a whole.

In advocating these processes to designers, supporting evidence can be drawn from a wide variety of component types and configurations which have to date been explored by most of the world's aerospace companies. It is recognised however, that these components have in the main been the subject of extensive and protracted development programmes based upon replacement of existing conventionally manufactured components and therefore have not been subject to the rigours of project timescales and programmes.

Nevertheless, these projects demonstrate suitable areas for the application of these technologies and provide a generic basis for the selection of appropriate components.

Ab-initio applications are now beginning to be applied on new projects and this is a certain indication of the achievement of maturity in the evolution of these processes.

Timescales for the development of simple components - in particular SPF - are relatively short, but for the more complex SPF/DB components, the timescale can be more protracted. This is due in the main to a current need to conduct an interactive design/manufacturing "trial and error" process to establish an optimum design and associated process route. The need for this pre-production development would be greatly reduced with the development of a Computer Aided Design (CAD) finite element analysis system. This would provide the designer with the capability of simulating the SPF process thereby providing him with a means of optimising the component design. Such analysis facilities would also provide a Computer Aided Manufacturing (CAM) system which will output the required pressure/time history for the respective component manufacture.

Because of their nature, it is the Authors view, that the design engineer needs to understand the details of these processes to a far greater degree than for most conventional manufacturing processes if he is to achieve his desired design objectives and more particularly in his selection of appropriate components.

A tradition has grown up over the years which limits the use of titanium to -

a) areas which demand its high temperature strength.

b) highly loaded, fatigue sensitive fittings.

This situation has arisen due, in the main, to the attendant high cost of material and associated conventional manufacture of titanium. Designers need to be reminded that the specific properties of titanium and its corrosion and fatigue resistance make it an attractive structural material even for room temperature use. With the development of the combined processes of SPF/DB and their cost saving potential, there is now a growing body of evidence that titanium can compete with aluminium fabricated components in both cost and weight. In the light of this evidence, designers therefore need to review their traditional views of this material.

High strength SPF aluminium alloys are now becoming available, and clearly these materials will find an increasing use in the manufacture of complex formed parts. Such parts are most likely to be associated with military and small civil aircraft, helicopters, and pylon/engine cowlings on larger civil and military transport aircraft, these being traditionally associated with areas posing extreme forming problems. DB of aluminium, currently does not provide the properties or ease of manufacture associated with titanium, but the flexibilities in design and the cost reduction potential provided by concurrent SPF/DB of high strength aluminium alloys remains a tantalising prospect for continued development.

Finally, although the main aerospace structural materials are aluminium and titanium, it should be noted by designers that a limited number of high temperature nickel alloys also exhibit superplastic characteristics. These together with the main aluminium and titanium SPF alloys are presented for reference in Table 1.

2.0 BACKGROUND

2.1 The Beginning

The major stimulus to the development of SPF and SPF/DB in aerospace component design and manufacture was the advent in the sixties of concepts for sustained supersonic cruise aircraft with Mach. Nos greater than 2.0. (Concorde, B1 and SST). These aircraft, with their high structural temperatures, demanded increased use of titanium in their primary structure. In recognition of this increasing demand, coupled also with the high cost of conventional manufacture of titanium structure, the Aerospace Companies who were primarily involved with the major supersonic cruise projects at that time - British Aircraft Corporation (Concorde), North American Rockwell (B1 - initiated programmes of manufacturing development which had as a primary objective the establishment of new manufacturing methods which would reduce the cost of manufacturing titanium components.

The characterisation of titanium alloys as superplastic materials by Backofen (1) in 1965 was timely and lead directly to the development and exploration of titanium superplastic forming as a potentially cost effective manufacturing technique.

Diffusion bonding had been used as a method of joining from ancient times and in the aerospace industry had been exploited for a considerable number of years in the cladding of materials for corrosion protection. DB as a manufacturing process in its own right was therefore being explored in the sixties as part of the new manufacturing development effort for low cost titanium component manufacture. By the early seventies the combined processes of SPF/DB for titanium structures had been established by British Aircraft Corporation and North American Rockwell (2)-(5).

Superplasticity had of course been recognised and demonstrated by metallurgists for a considerable number of years prior to the sixties, but these demonstrations were mainly associated with materials and alloys which were of little interest in aerospace structural design. Some aluminium alloy developments were however carried out in the sixties by the then British Aluminium Co. (BACO) which were the forerunner of the present day Superal 100 and 150. These alloys were initially formed by the then British Aircraft Corporation, but because of their relatively low strength, only limited development was undertaken at that time, preference being given to the development of SPF and SPF/DB of titanium.

2.2 Development Phase

The decade from the early seventies to the early eighties has been one in which the SPF and SPF/DB processes have developed from the small laboratory scale basic process demonstrations, to full size component design, manufacture, structural and flight testing. This decade has also seen the spread of the technology to all of the worlds major aerospace companies.

The largest programmes of development have been conducted in the USA involving substantial government funding which was reported in 1984 (R6) as being in excess of $23m. These programmes have involved most of the US major aerospace companies. Parallel developments have also taken place in the UK since the late sixties in particular involving British Aerospace (titanium and aluminium) and Superform (aluminium). Government sponsorship for this activity in British Aerospace has also been provided in the UK. More recently work has also been reported from Germany, France and Japan (7)-(10). Although not intended to be an exhaustive list, Tables 2 and 3 provide a summary of the most significant development components associated with these programmes and which are reported in the literature. Comments are also made in Tables 2 and 3 with respect to salient structural or design information associated with each component.

The primary objectives of these programmes have been to demonstrate -

a) the cost and weight advantage provided by SPF and SPF/DB.

b) the practicability of SPF/DB in the manufacture and design of full size components.

These programmes show a consistency in their conclusions and indicate positive results with respect to their objectives. A summary of the advantages derived from these programmes and which now seem to have a universal acceptance is given in Table 4.

Apart from the demonstration of the attainment of the objectives, it is of interest to note from a design point of view the wide range of component types that have been considered in these programmes and the range of loading and test conditions which have been successfully applied. More particularly, however, it is important to note the number of components for which titanium SPF/DB provides cost and weight advantages over the equivalent component conventionally fabricated in aluminium. This latter fact is perhaps the most significant from the point of view of the level of embodiment of SPF/DB titanium that might be anticipated on future aircraft.

A further significant factor which is in favour of titanium SPF/DB and one which is gaining in prominence in our increasingly cost conscious environment is the favourable "cost of ownership" or "life cycle cost" which is associated with the corrosion resistance and fatigue qualities of titanium. These claims are made in particular in respect to the replacement of aluminium honeycomb as in the case of the T-38 main landing gear door developed by McDonnell/Douglas. (7).

During this period of major development in titanium SPF/DB, work has also been proceeding in more recent years with the development of high strength SPF aluminium alloys and other high strength alloys which are of interest to the aerospace designer (Ref. Table 1).

Cavitation and its effects on material properties has been a fundamental problem inhibiting the successful exploitation of high strength aluminium. It is significant therefore that in parallel with the improvement in the strength of aluminium alloys, important work has also been proceeding with the development of methods of cavitation suppression to maintain the properties of these alloys following SPF. The development of these alloys and the effects of cavitation are discussed and presented in detail by Dr. Grimes and Dr. Ridley in their respective papers presented in this series. The results of these developments have now reached the stage that an early exploitation of high strength SPF aluminium on production aircraft is likely in the near future.

Whilst DB of aluminium has now been demonstrated to be technically possible,-without the need to subject the material to the high strains which are associated with the bonding achieved in the already proven cladding process,-the possibility of a cost effective combined SPF/DB process with the full range of structural possibilities provided by titanium has yet to be demonstrated. Distortion free heat treatment for property recovery, and corrosion protection of such aluminium structures will also need to be addressed. Properties of aluminium DB although significantly higher than conventional aluminium joining techniques such as riveting and adhesive bonding have yet to achieve the parent strength and consistency already demonstrated by titanium. These aspects are addressed by Dr. Partridge in his presentation in this lecture series.

In parallel with the developments in materials, processes and demonstrator components important developments have inevitably taken place in the associated manufacturing equipment and technique. Early experimentation was conducted using primitive press/heated platen systems and in many cases using "hard back" bolted up tooling systems. Today, however, purpose built heated platen presses of a type similar to that shown in figure 1. have become the mainstay of SPF and SPF/DB manufacture.

Variations on this basic theme have now been produced in, for example, the "Shuttle" platen press developed by Aerospatiale, the Gantry press system recently installed by McDonnell/Douglas and the "C-Frame" restraint system developed by Grumman. Specialised equipment has been developed for SPF production, for example, the patented press developed by Superform for aluminium the principles of its operating are presented in figure 2, the "infra red" heating system developed by McDonnell/Douglas and the "HOT box" principles developed by Rockwell, figure 3. Techniques for "thinning" reduction have been developed, for example, the principles associated with the Superform press, the use of a separate forming membrane, figure 4, and the use of reduced friction tool coatings. The successful development of such manufacturing equipment and techniques are clearly a key factor in the maximisation of the cost benefits provided by these processes and are a certain indication of the growing success of these techniques. In the Authors view further generations of new and more efficient production systems are likely.

2.3 Production Phase

The period from 1980 onwards has seen the increasing exploitation of SPF in the manufacture of production components in both the USA and UK. It has been reported (6), that some 230 titanium parts were in production by 1984 on F-15, F-18, AV8B and B1. In the UK in 1981 4 primary structural SPF parts had been introduced for the first time onto a civil aircraft Airbus A310, figure 5 and 12 parts had been introduced onto Tornado by 1983. SPF/DB production parts have now been successfully introduced onto B1, Airbus A310, A320 and Tornado.

The number of SPF/DB parts in production has recently greatly increased as a result of the redesign of the F15 rear fuselage/nacelle area to take account of the advantages offered by these processes. Although the applications referenced above have all been replacements for existing conventionally manufactured components, ab-initio applications have now been introduced onto the current Agile Fighter aircraft in Europe - British Aerospace EAP (keel member) and Marcell Dassault Rafale (leading edge slats). It is of significance to note that the Airbus and Rafale titanium SPF/DB components are replacements for what would traditionally be aluminium conventionally manufactured equivalents.

Further production components are a likely possibility in the near future based upon the development components listed in Tables 2 and 3. The T-38 undercarriage door which has already been the subject of a limited production run is reported as being a likely candidate for further production - again this represents titanium SPF/DB replacing conventionally manufactured aluminium. Although not employing SPF, it is nevertheless highly significant from the point of view of confidence in the ability of these processes to produce production components to meet the widest range of demanding conditions, that Rolls Royce have introduced DB into the manufacture of their wide chord fan blade for their RB211-535E4 engine. Further developments of the application of titanium SPF/DB to engine components (blades in particular) is now being actively pursued (11) and represents an area of application with huge production potential.

Production of a range of SPF parts by Superform in Supral 100 and 150 has taken place over a number of years (12) and it is anticipated that the advent of high strength aluminium alloys will greatly increase the number of production SPF components in the near future.

In conclusion therefore the pioneers of these processes can be satisfied that within two decades, which is the normal gestation period for a new aerospace technology, a range of production components are now being successfully produced which exploit SPF and SPF/DB, and that there is every indication that these technologies will continue to grow in significance. It is interesting to speculate however that full scale production could have been a realistic possibility some ten years earlier but for the fortunes of projects such as B1 and Concorde which gave the initial impetus to these developments.

3.0 SUPERPLASTIC FORMING (SPF)

Superplastic sheet forming has been described by Dr. Hamilton in his earlier paper in this lecture series and is a manufacturing process which utilises the characteristics of SP materials to undergo elongations of several hundred percent without necking or local failure. The process is associated with temperatures in the region of half melting point and is sensitive to strain rate, figure 6. At the optimum forming temperature and straining conditions, rates equivalent to some 100% strain in one hour are typical, therefore, for complex shapes with fine detail and small radii the total forming time can be several hours.

Superplasticity, its mechanisms and characteristics are fully described by my colleagues and the characteristic relationships of flow stress, strain rate and 'm' values as illustrated in figure 7 have by now become very familiar.

The process route which exploits these characteristics is shown in figure 8. and as can be seen this involves primarily the inducement of a membrane stress in the material being formed by applying a gas pressure, whilst maintaining the forming material at the appropriate superplastic temperature. This basic process applies to all materials but with certain detail differences for particular materials. For example, because of the high temperatures associated with titanium forming and the reactivity of this material at this temperature, it is necessary to use high purity argon gas in the forming of this material.

In the case of aluminium alloys, although reactivity can be discounted, cavitation, as has been previously described, is an attendant undesirable feature of aluminium forming. This problem can however be suppressed by forming with a high ambient background pressure onto which the forming pressures are superimposed.

Under the action of the <u>local</u> induced stresses the material achieves a <u>local</u> strain rate which is governed by the material characteristics shown in figure 7. As these <u>local</u> stresses are a function of <u>local</u> thicknesses and curvatures, both of which vary during the transformation from the initial blank to the finished formed component, it follows therefore that to maintain the optimum forming conditions throughout, it is necessary to apply a variable pressure time history which is unique to each component. It also follows that as the process of material movement is a function of local conditions it is not possible to drive all of the material at the maximum 'm' conditions using a uniform pressure system. A compromise therefore needs to be reached if a practical and simple forming process is to be arrived at. This usually means driving the fastest forming element at the optimum 'm' with all other elements of the component forming at the lower local rate that results from the pressures applied.

Relative to conventional forming processes, the strain limits associated with SPF are one to two orders greater. This is illustrated in figure 9. which shows the unique relationship of elongation to failure as a function of 'm'.

Superplasticity is usually defined as having an 'm' value greater than 0.25, but this is however low by comparison with the normal range of 0.5-0.8 that would be associated with materials of interest to the aerospace designer. In most applications of SPF, the strain limits do not usually exceed a level much greater than 300%. Hence as can be seen from figure 9. even at a relatively low 'm' value of 0.5 the material has almost double this strain capability before failure. These extreme strain capabilities and the available margins lead to the primary advantages associated with SPF which are its ability to produce complex shapes in <u>one</u> operation which far exceeds that possible with conventional forming processes. Furthermore, the growing body of evidence from volume production supports a claim of exceptional accuracy and repeatability for the SPF process.

4.0 SPF DESIGN ASPECTS

The primary design and manufacturing objectives associated with the selection of any processing route and associated material are the achievement of:-

a) maximum structural efficiency
b) minimum cost
c) a compromise based upon a) and b).

As has been stated previously, we live in an increasingly cost conscious world and hence b) and c) are gaining in significance as primary design considerations. The generally accepted benefits provided by SPF in relation to these objectives are presented in Table 5 and within the limitations of their respective mechanical properties these advantages are equally applicable to all of the SPF materials which are currently available to the aerospace designer.

We shall now consider each of the design objectives in turn and review the extent to which SPF can satisfy these objectives.

4.1 Structural Efficiency (Minimum Weight)

The main structural efficiency advantages provided by SPF can be categorised under three headings all of which have as a basis the superior shape making capabilities associated with the SPF process.

These three categories are:-

a) shape or form which improves structural stability and therefore load carrying capacity in compression and shear carrying structure.

b) shape or form which reduces joints thus saving weight penalties which would otherwise be associated with overlaps and stress raisers.

c) consistent shape and finished thickness (including for SPF distribution of thickness) which therefore minimises variability in structural performances in a given design.

4.1.1 Shape for Structural Stability

Structural designers are aware of the advantages provided by corrugated/swaged structures and already exploit these advantages in existing designs. These applications are of course conditioned by the capabilities and limitations of conventional manufacturing technology to produce cost effective shapes. In airframe design, these structural features are currently exploited to a lesser degree than perhaps is found in space vehicle structures for which corrugation is a common feature. This perhaps reflects the relative premiums that are acceptable for weight saving in these respective areas.

In support of the added structural advantages that can result directly from the extended shape making capabilities of SPF, we can draw upon data published in the literature. This data applies to the three areas for which such features are of significance - compression panels, struts and shear webs. In all three areas considerable weight advantages are predicted and demonstrated for SPF, but it should not be forgotten that in designing such optimum structures, failure under load can be sudden and is also associated with a much reduced post buckling capability. Such structural behaviour may be less attractive to the Civil Airframe designer for whom crashworthiness may be an important design consideration. The degree of optimisation which is tolerable in any design must also take into account practical limitations such as:-

a) degradation in performance due to in service damage for long life structures.

b) the need to facilitate the attachment and routing of aircraft systems and controls.

c) the limitations in consistency and control of shape and thickness afforded by the manufacturing process.

All of these considerations can have a significant effect on the performance of structures which are stability designed. The latter consideration, as it relates to SPF is discussed in a later section of this paper.

a) Compression Panels

The optimisation of compression panel structures which are suitable for manufacture by conventional processes has been studied in depth by Emero and Spunt (13). Their analysis involves the derivation of an efficiency factor which provides a basis for ranking the various configurations. Figure 10 provides a summary of this analysis and indicates the maximum efficiency ascribed to each configuration. This demonstrates that the familiar trapezoidal corrugated sheet represents the structural form which exhibits the maximum efficiency. Using this basic shape, Davis etal (14) have explored the possibility of further improvements in structural efficiency by adding shape to the webs and caps. The various configurations considered in their analysis are presented in figure 11 and the resultant improvements in efficiency are presented in figure 12. Clearly such configurations are only achievable by the use of SPF and a combination of SPF and DB.

The Author is not aware however of a practical application of such a structural configuration, but the theoretical benefits predicted suggests that the manufacture and testing of such a configuration is worthy of consideration.

b) Compression Struts

Because of conventional manufacturing constraints, compression struts are in the main usually of a plain constant section tubular form with in some cases simple tapers to reduce the end fitting weight.

The substitution of an SP material enables additional shape and hence improved performance to be achieved by inflating the basic tube into a tooling system (15). In addition to local stability features, SPF also enables the cross section of the stabilised tube to be varied along its length in such a manner as to improve Euler capability. Such a strut manufactured in titanium 6AL-4V is illustrated in figure 13 and the theoretical predictions of the improvements relative to plain tubes are shown in figure 14 for both aluminium and titanium alloys.

As can be seen, considerable weight advantage can be obtained by the exploitation of the SPF process in the manufacture of optimum compression strut configurations.

c) Shear Webs

The use of corrugated shear web structure has been explored in the SP manufacture of major structural items such as nacelle frames for the B1 aircraft (16) under US Air Force Materials Laboratory contracts. These exercises demonstrated weight savings of between 33 and 40% for the titanium SPF structures relative to their conventionally fabricated titanium equivalents.

In conclusion therefore it is clear from the previous three sections that SPF can provide significant improvements in the structural efficiency of stability designed structure with weight savings up to 40% being attainable.

Figures 15 and 16 provide details of the location and shape of an Airbus A310 component which has now been in SPF production since 1981. This component is a primary structural item which is stability designed - in the event of a crash with full fuel tanks, it is required to withstand an implosive pressure of some 70 psi (5 bars). The superior shape achieved by SPF can be contrasted with the best shape achievable by conventional means which is illustrated in figure 17.

This difference in shape has lead directly to a saving in weight of 35% for the SPF component. This component is believed to be the first primary structural SPF application on a civil aircraft. The Airbus A310 fleet time at the beginning of 1987 had achieved in excess of 400,000 hours with the lead aircraft having achieved 10,000. No adverse experience has to date been reported for these particular components.

4.1.2 Shape to Reduce Joints

Because of the limitations of conventional processes to achieve certain shapes, it is frequently necessary to manufacture components from a number of separately formed parts which are then subsequently joined by conventional means (welding, riveting).

Because of the property reductions associated with joints and the need to provide overlaps to create such joints there is inevitably an associated weight penalty. The vastly improved formability of SPF material enables many of these joints to be eliminated and hence a weight saving can result. Figure 18. provides a simple demonstration of the use of this principle applied in practice. The component shown represents a Concorde panel located in the engine bay doors manufactured by conventional processing and by SPF. The elimination of joints which was facilitated by SPF of titanium resulted in a weight saving of 20%. Northrop Corporation report (17) the exploitation of this principle in the redesign and manufacture of the ejector valve bracket illustrated in figure 19. This demonstrates very well the need for the designer to have a detailed awareness of added flexibilities provided by the SPF process if he is to achieve and exploit its benefits to the full.

4.1.3 Consistency of Shape and Thickness

In the optimisation of his structural design, the designer needs to make allowances for the variability in structural performance which will arise due to the dimensional limits which are achieved by the manufacturing route which is adopted.

For SPF, a considerable volume of evidence is now becoming available from components which are in full scale production. This evidence supports the claim of exceptional consistency and accuracy for these processes. In the Authors experience and based upon the manufacture of the Airbus components illustrated in figures (15) and (20), for which some 700 and 2000 components had been produced respectively by early 1987, a tolerance of +0.010" is achievable on a 2 ft dimension. The tolerance limits quoted above represents ± two standard deviations derived from measured dimensions taken from these components. In addition, an analysis of thickness measurements taken at a significant number of points on these components indicates that formed thicknesses are consistent with the basic starting sheet tolerances (for titanium ± .004") and although SPF gives rise to thickness variations across the component-which are primarily a function of its shape-thickness at corresponding points on components can be maintained to within the tolerances stated above.

In conclusion, therefore, and based upon evidence from significant production runs, the structural performance of SPF components should be subject to low variability from the shape and thickness achie ed after forming. This together with the versatile shape making capability of SPF, make this process attractive for the manufacture of optimum stability designed structure.

4.2 Design Aspects in Relation to Cost

Table 5 provides a summary of the potential cost benefits associated with SPF in relation to the basic characteristics of the process. It is of interest to note that there is a potential cost benefit associated with each of these characteristics. We shall now expand these summary statements with a view to highlighting the designers role in the maximisation of the cost benefits associated with his selection and design of an SPF component.

4.2.1 Forming Capability in Excess of Conventional Forming

The components illustrated in figure 18 and 19 provide simple and effective examples of the use of the superior formability of SP materials. The reduction in the "parts count" achieved in the design and manufacture of the SPF components has obvious cost advantages as well as the weight advantages previously referred to.

In designing such components, it must be remembered that, although SPF can achieve extremely fine formed details, the process is strain rate sensitive and consequently, the designer should incorporate the maximum corner radii possible, consistent with the design requirements for the particular component under consideration.

Figure 22 illustrates the effect on forming time and corner thickness of various corner radii. These effects are defined relative to a radius of 0.5". As can be seen, the time and hence the cost penalties which can result from the requirement for small corner radii can be considerable. A further cost consideration arises from the reduction in thickness associated with corner radii. This relates to the fact that as such areas can be critical from the point of view of structural integrity, a minimum thickness will be required in a given design. This in turn will dictate the initial sheet thickness required for the manufacture of the component and or the need for selective reinforcement of the blank prior to forming both of which have cost implications which can be influenced by the design.

4.2.2 "One Shot" Process

In conventional forming, it is frequently necessary to carry out a number of separate pressing operations to achieve a desired shape. These are usually preceeded by intermediate heat treatment stages which can lead to a protracted forming operation. In addition to the time involved in these operations such a manufacturing route can also require the provision and use of a number of intermediate tools. The cost implications of both of these factors relative to a "one shot" process, (all be it a strain rate sensitive process) are obvious.

As discussed in the previous section, the duration of the single SPF operation is related directly to the detail requirements of the particular design. The exploitation of the single SPF operation is a significant factor in the cost saving of 35% achieved in the manufacture of the component illustrated in figure 16. The conventionally manufactured equivalent illustrated in figure 17 required some eight passes with intermediate heat treatment. Even with this protracted process, the final shape failed to achieve the same structural definition as is possible with SPF. Clearly in designing the SPF equivalent for this component, the designer compromised some of the cost saving potential to achieve a reduction in weight, by demanding a more severe shape in the SPF component than was achievable in the conventionally manufactured equivalent.

4.2.3 Accuracy and Repeatability

The cost implications for these particular attributes of the SPF process are a potential to reduce assembly and fitting times. In addition, they provide the designer with the choice of a formed part which could provide accuracies which are akin to a machined alternative.

4.2.4 Simple Cavity Tooling

Whilst the primary cost advantages of such a feature is more directly associated with the reduced cost of the tooling required (relative for example to matched tooling), the designer can exploit this feature of SPF by blowing a complete envelope from a simple blank or preform, rather than producing separate half pressings which are subsequently joined. This method was adopted for the SPF manufacture of the component illustrated in figure 16. Apart from the reduced forming time associated with this technique, accurate control was also maintained for the envelope cross section which in turn had a beneficial cost effect on the fitting of the mounting flange which was attached to the component by welding after forming.

Although re-entrant features can be formed in an SPF component, such features should be avoided by the designer because of the added complication required in the design of the tools to facilitate withdrawal of the component. This added complication obviously increases the cost of manufacture of the tools. A further undesirable feature of the re-entrants is an adverse effect on material thinning which further reinforces the need to avoid such features in component design.

To facilitate the removal of a component from an SPF tool, parallel sections should be avoided in the component design and drawn angles of between 3° and 5° should be provided wherever possible.

4.2.5 Negligible Tool Wear

Although not a fundamental aspect that needs to be taken into consideration by the designer in the derivation of his component design, the fact that SPF is associated with negligible tool wear is a significant cost factor in support of the SPF process. Practical evidence in support of this claim is increasingly becoming available. In the case of the component shown in figure 16, some 300 components have now been produced in one tool without degradation. The dimensional accuracies achieved in the manufacture of this component and reported in an earlier section of this paper is evidence of the tooling stability which is currently being achieved.

4.2.6 Process Times Independent of Size or Number

SPF is a strain rate sensitive process and therefore the process time associated with the forming of a component is primarily a function of the maximum strain (or minimum thickness) that is experienced in the forming, the processing time is to a first order independent of the relative sizes of the components or alternatively the number of the same size components which are processed at the same time.

Different components can be grouped together and formed in a composite tool thus maximising the use of the available heated platen area and minimising the cost of forming. As each different component has its own unique forming cycle which is specifically designed to achieve optimum conditions, it will be necessary for the designer to compromise on these individual optimum cycles to achieve a common cycle which can be applied to the group of components. The implications of applying such a cycle will be more adverse thinning on certain components but this effect can be minimised by the grouping together of components which have similar optimum cycles.

In conclusion therefore it is clear that there are a considerable number of cost factors which need to be addressed by the designer and which require an in depth understanding of the basic process details.

4.3 Thinning

The essential driving mechanism of SPF is the inducement of a flow stress in the sheet material by means of a gas pressure, as illustrated in figure 8. In the majority of components, the stress levels are rarely uniform due to local curvatures of the forming sheet, edge constraints and friction associated with local tool contacts. This variability in stress produces variable forming rates across the component and therefore results in variable thicknesses in the finished component.

For a given component, and controlled processing conditions and using material with similar SP characteristics, the thickness variations in finished components are repeatable and predictable to within close tolerance as discussed previously.

From a design point of view and if required, the thinning effect can be compensated for by:-

a) the use of variable thickness starting blanks (figure 25).

b) male/female forming (figure 26).

c) preforming (figure 2, figure 24)

d) selective reinforcement (use of DB).

e) the use of reduced friction coatings.

Adjustments to thicknesses on the finished components can also be made by varying the process conditions. This effect is illustrated in figure 27 which shows the result of processing at different 'm' values on the thickness distribution in a blown hemisphere, which is the basis of all SPF component manufacture. 'm' values are a function of temperature as illustrated in figure 25, but apart from using temperature as a possible means of thickness control, this fact emphasizes the need to control and maintain uniform or consistent temperatures in the manufacture of a component if the quality of the finished component is to be preserved. Consistency in material SP quality is also an important factor in maintaining consistency in the finished component. Usually the material is qualified on the basis of micro-structure with a follow up forming test for material for which the micro structure examination is inconclusive. This procedure has been adopted by British Aerospace and has proved successful for several thousand production parts which have been produced to date. With the increasing application of SPF components it seems essential that a specification should be drawn up which defines the quality standards required by the manufacture in the material supply.

4.4 Quality Assurance

Apart from the basic SP material quality aspects discussed in the previous section, there are in the Authors view, no fundamentally new quality assurance aspects which need to be addressed for the SPF production of titanium parts. In the case of aluminium SPF components, reliability and consistency of cavitation suppression technique will need to be addressed.

Although thickness variations in SPF components are an inherent feature of this process, production experience has shown that the consistency in thickness distribution between components is such that thickness measurements can in practice be limited to a small number of key monitoring points.

4.5 Application of SPF

A number of examples of the successful application of SPF have been sited in the earlier sections of this document. These represent generic types of application and in summary these are -

Stability designed structures

- ribs
- frames
- beams
- compression struts

Complex multi element sheet components

- panels
- mounting brackets and supports

Complex envelopes

- ducting
- tanks
- vessels

Decorative panels and furnishings

It is considered appropriate at this juncture to re-emphasize the need for the designer to establish an SPF design which makes full use of the process capabilities, if the advantage in cost and weight afforded by these processes are to be realised. Although the Author is not aware at this point in time, of a component which solely employs SPF of titanium as a replacement for the conventional manufacture of an aluminium component, this possibility should not be ruled out by the designer.

5.0 Diffusion Bonding

DB has been exploited as a method of joining from ancient times but in aerospace manufacture its earliest use was associated with cladding of materials for corrosion protection. This process is generally carried out in cold/high pressure conditions associated with heavy rolling and also involves significant straining/deformation of the materials to be bonded.

By contrast the latest development in the DB process are associated with high temperatures and relatively low pressure with low or negligible strains.

Two categories of DB are recognised:- These are "solid state" or "liquid phase". From a practical point of view, the primary advantages of liquid phase relative to solid state bonding is the reduced pressures and times associated with liquid phase bonding.

5.1 Theoretical Predictions

The early empirically derived processing conditions for DB, have now been supported by the establishment of a number of theoretical models of the process (18)-(23). These models vary in their assumptions but in all cases demonstrate the dominance of surface topography in the determination of the processing conditions required to achieve a bond.

For solid state bonding, the models consider surface diffusional mechanisms for areas of intimate contact, preceeded by deformation mechanisms such as power law creep and plastic yielding.

Work has also been done on the modelling of liquid phase bonding when using low melting point interlayer between the bond faces (22). The primary advantage of the interlayer is its ability to "wet" the bond surfaces and effect "erosion" type mechanisms coupled with rapid transport within the liquid. This model confirms the practical experience of reduced pressures and times to effect a liquid phase bond.

5.2 Process Conditions

The main processing variables associated with DB are:-

- Temperature
- Pressure
- Time

5.2.1 Temperature

All of the mechanisms associated with DB - diffusion, creep, plastic yielding etc. are assisted by carrying out the DB processing at high temperature.

In a practical situation, it is usual to maintain the temperature of the process constant and uniform and as high as possible subject only to material considerations, such as alloy phase transformation temperatures, and liquid phase temperatures. By maintaining the temperature as high as possible the time to effect a bond is minimised for any given bond pressure and surface condition.

5.2.2 Pressure

Because of the creep and plastic yield mechanisms, pressure is an important process parameter. Clearly the higher the pressure that can be exerted at a given temperature the lower the time to effect a bond. The level of pressure employed however, is usually a compromise between the economic considerations of time, capital equipment, and tooling material, the latter being a particular consideration for the high temperature conditions associated with titanium DB.

As the "wetting" effect associated with liquid phase DB brings the two joint faces into direct contact without the deformation required in solid state bonding, the pressures associated with liquid phase bonding are reduced.

5.2.3 Time

Apart from the obvious economic consideration associated with the time to effect a good quality bond, consideration also needs to be given to time dependent metallurgical phenomena such as grain growth in the material being bonded and the consequent reduction in mechanical properties that might result. In addition, consideration needs to be given to the effect of grain growth on the SPF performance of the base material if the combined SPF/DB process is being used (24). In general however, these metallurgical effects are minor and enable processing times of several hours to be accommodated without significant degradation in material properties.

5.3 Material Requirements

The essential material requirements for DB relate to conditions of the mating faces to be joined. These are:-

1. Flatness
2. Surface roughness
3. The absence of insoluable substances on or between the faces

Because of the essential requirements of intimate contact between mating faces as a prerequisite for bonding by atomic diffusion, the surfaces to be joined should be as flat and smooth as possible. For practical applications of DB it is usual to specify a roughness equivalent to a ground surface (RA Circa 0.5 mm). Rougher surfaces can be accommodated however, by appropriate adjustments in the process conditions. Figure 28 shows for reference the surface roughness associated with a range of surface machining operations.

Although bonding times can be reduced by the use of the finest possible finish, this needs to be balanced against the time that would be required to achieve such a finish.

The existence of insoluable substances on or between the surfaces will prevent bonding. These "no bond" areas manifest themselves as voids in the case of entrapment of insoluable gasses such as argon, or alternatively areas of intimate contact in the case of insoluable surface layers such as oxide layers.

The inhibition of bonding by the introduction of insoluable surface layers has a practical significance in the manufacturing process, as it enables selective bonding to be carried out and thus facilitates the manufacture of highly complex structural forms by the combined SPF/DB processes. This important aspect is discussed later in this paper.

5.4 Titanium as a bondable material

Because of its associated superplastic properties the 6% Aluminium 4% Vanadium titanium alloy has become the most widely used and recognised high strength titanium alloy in the manufacture of DB components. DB temperatures for this alloy are usually in the region 930-950° C - this temperature range being conditioned/limited by the phase transformation temperature for this two phase alloy of 985° C.

At temperatures in excess of 800° C titanium and its alloys become highly reactive and will absorb their own limited oxide layer, thus producing a "self-cleaning" action which is compatible with the DB process requirements. The reactivity of titanium at these temperatures does however require that the material is protected in an inert (argon) atmosphere or alternatively under vacuum. Because of the insoluability of argon in titanium, care needs to be taken, however, to ensure that during the bonding sequence the argon is expelled from between the mating faces and does not become entrapped, thus producing large voids in the bonds..

Because of the self-cleaning action associated with titanium alloys, these materials can be used in the bonding process as "as received" materials with the pre-preparation of sheet for example being limited to standard cleaning processes such as degreasing and "pickling" (acid etch). This factor obviously enhances the economic advantages of the use of titanium in low cost component manufacture.

Although solid state DB of titanium is readily achieveable, there are practical advantages to be had in the use of liquid phase DB. In particular, the use of an interlayer can significantly reduce the pressures required to effect a bond. Fitzpatrick (25) reports that the use of a Cu/Ni interlayer, electro plated onto the bond faces, allows bonding pressures to reduce to less than 700 mb.

This contrasts with sheet titanium solid state bonding pressures of 20 to 30 bar which are typical of current component manufacture. Based upon the familiar Wiesert/Stacher relationship presented on Figure 29 the associated times to achieve a bond are 1 to 1.5 hours.

5.5 Aluminium as a Bondable Material

The major obstacle which prevents aluminium alloys as "as received" materials being used for DB, is the existence of a tenacious oxide layer which is insoluable even at DB temperatures. Before aluminium and its alloys can be bonded therefore, it is necessary to either remove this oxide layer by chemical/mechanical cleaning, or to substitute a soluable alternative which will be absorbed by the base material at the DB temperatures - thus behaving in a self cleaning manner akin to that for titanium. The introduction of a substitute layer can also be used as a means of effecting liquid phase bonding with a resultant reduction in bonding pressures and times.

DB of aluminium in the form of cladding has been in use for a considerable number of years. This process involves cold/high pressure rolling and considerable straining of the material, which causes break up of any residual oxide layer.

The pretreatment of the aluminium alloy involves chemical/mechanical removal of the oxide layer. Because of the capital equipment involved this bonding technique has not been viewed as a method for component manufacture but recent developments by Texas Instruments in the manufacture of components by their "thermally expanded metal" process (26) could result in cold roll bonding gaining some prominence in future aerospace component manufacture.

For low pressure/high temperature bonding, which is being developed in association with SPF and MMC fabrication, the method of oxide pre-treatment involves either chemical/mechanical cleaning or alternatively the substitution of the oxide layer with a thin absorbable metallic coating of elements such as Cu, Ag etc., the choice of element being subject to the base alloy being bonded.

Coating techniques such as Ion Vapour Deposition (IVD) allows these absorbable coatings to be of minimal thickness and hence their absorption has negligible effect on the base alloy composition and its associated mechanical properties.

Development work carried out by British Aerospace concludes that the oxide layer substitution technique provides higher quality bonds with less variability in bond strength.

DB temperatures for aluminium alloys are in the region of 500° C, but the actual temperature used is subject to the base alloy and the metallic coating employed.

Bonding Pressures for aluminium alloys using the "hot" DB process are relatively low compared with the roll bonding process and times as low as 15 minutes have been achieved.

5.6 Bond strengths

The following table summarises typical strengths of DB joints in titanium and aluminium alloys. These are compared for reference with parent material strengths and typical rivetted and adhesively bonded joint strengths.

Mechanical Properties of DB Joints

Material		Shear Strength MPa
Titanium 6 AL. 4V	Parent	575
	DB	575
Aluminium	Parent	320
	DB	150 - 170 (27)
Riveted Joint (Typical)		10
Adhesive Bond Joint (Typical)		20 - 40

What is evident from this table is the capability of titanium to achieve parent strength properties and the significant increase in shear strength of titanium and aluminium DB joints relative to current conventional joining techniques.

Baker and Partridge (28) have demonstrated that for good quality bonds in titanium the fatigue performance of DB joints is equal to parent material but that the existence of micro voids in the joint area will degrade the fatigue performance without significantly affecting the static strength. Work by the same authors has demonstrated that even for joints with minimal voiding a significant reduction in impact strength is experienced at a DB joint in titanium.

From a component design point of view, parent material fatigue strengths are usually of academic interest as these properties are usually degraded to take account of stress concentration effects associated with notches and joints. For the case of titanium 6/4 the parent material endurance limit (10^7 cycles) of 530 MPa would be reduced to 330 MPa for normal design purposes i.e., 38% reduction. Data from the literature and from testing conducted by British Aerospace, suggests that this reduction in parent strength represents a realistic degradation factor to cover practical limits of micro voiding.

The effects of micro voids established by Baker and Partridge are presented in figures 31 and 32.

A survey of the literature by the Author indicates that by comparison with the volume of research reported on the SPF mechanisms only limited work is reported in the field of DB in particular in respect to bond failure modes when associated with defects and in the development of an understanding of the reasons for low impact strength of DB joints and a method of improving the joint properties.

5.7 DB Joint Defects

DB joint defects fall into three categories. These are:-

1. Micro voids
2. Large voids
3. Intimate contact disbonds

Figure 32 shows a typical micro void in a DB joint - such a defect is only likely to occur as a result of the application of the incorrect processing conditions. This defect can be largely overcome by ensuring that the processing conditions of pressure and or time are chosen to encompass the widest range of surface conditions likely to be encountered in the manufacture of a given component.

The effect of the presence of micro voids has and continues to be studied. Results so far indicate some insensitivity to micro voids in static strength properties but reductions in fatigue, impact and elongation to failure have been demonstrated.

The cause of large voids/disbonds is most likely to be associated with argon gas entrapment when bonding in an argon environment. This defect can be overcome by employing the correct processing technique and in the detail design of the component to ensure progressive venting of the faces to be bonded. Large void areas can also arise due to irregularities in the platens or workpiece when carrying out "massive" DB. In this case, the bond faces have failed to come into contact because the pressures employed are generally insufficient to forge the material into the required shape. The effect of large voids on component integrity is currently being studied - of particular concern is whether such voids, even if deemed to be initially non critical because of their location on a component, will propagate to a critical state in a service environment.

As discussed previously, intimate contact disbonds are associated with non gaseous insoluable layers between mating faces. These insoluable areas can either be present on the faces to be bonded, or arise during manufacture due to inadequate pre-processing or as a result of the ingress of surface contaminants during the DB processing. The prevention of such disbonds is clearly dependent upon good process control. As in the case of large voids, such disbonds may not be critical on a particular component, provided they do not propagate, but unlike large voids the intimate contact in these disbonds makes their extent difficult to ascertain by current NDT techniques.

5.8 N.D.T.

Of the three defect types discussed in the previous section, only the large voids can be readily detected by current N.D.T. techniques such as C-Scan ultrasonics and X-Rays (29)-(30).

Detection of micro voids can only be accomplished by micrographic examination and in a practical situation this is usually carried out on a sample basis either by cut up of production components or by the examination of areas of the components which are removed during subsequent production operations; control of such defects is however best accommodated by correct processing and process control.

Recently published information (31) suggests that such defects may be detectable by the use of scanning acoustic microscopy. This technique relies upon the fact that it is possible to focus an acoustic beam to a diffraction limited spot by making use of the large velocity mismatch between sound waves in material such as sapphire and water. An S.A.M. lens based upon this principle is shown on figure 33. As reported such devices are capable of measuring cracks and discontinuities $<$ 100 n.m.

Intimate contact disbonds is a defect which is common to all bonding processes and is therefore not unique to DB. Detection of such defects presents difficulties for all bonding processes and this is currently the subject of a considerable amount of research using techniques such as thermography, holography etc. An interesting development in this field is the use of a YAG laser as a means of inducing a temperature differential and therefore distortion of the unbonded laminate. The distorted shape is then detected by a focussed interferometer (32). The principle of this technique is also illustrated in figure 33.

As in the case of micro voids and in common with other bonding processes currently accepted for aircraft production, close control of the processing route is seen as the most practical solution to this particular problem.

5.9 Component Manufacture

5.9.1 Equipment

The process of DB requires a means of maintaining the workpiece at a uniform temperature, together with the means of applying and reacting the loads required to produce intimate contact between the faces to be bonded. The equipment most widely used as a means of providing these conditions is a heated platen press, a typical example of which is shown on Figure 1. Provision of a protective environment in particular when DB'ing titanium is either provided by the addition of a vacuum chamber as an integral part of the press or alternatively as an integral part of the individual component tooling. In the case of certain SPF/DB forms the DB is protected within the titanium blank.

5.9.2 DB Processing

DB in component manufacture is generally categorised into two forms:-

1. "Massive DB"
2. "Thin Sheet DB"

Massive DB involves the joining of machined plate elements and is generally associated with the manufacture of thick section components which would otherwise be machined from solid, or alternatively, from a forged blank. Figure 34.

As a general rule, the bonding pressures associated with "massive" DB are applied by mechanical means and can require the application of loads in more than one direction.

The DB joints in components manufactured by this route are a substitute for parent metal in the conventional equivalent and would in general be subject to high levels of stress in service use; this in turn demands high integrity bonds in the manufacture of these components if good structural efficiency is to be maintained in the DB component.

The need to achieve such high quality bonds coupled with the remoteness of the applied mechanical bonding loads places considerable demands on the accuracy required of both the tooling and the individual plate elements employed in the component manufacture.

By contrast the use of "thin sheet" DB has two practical advantages relative to "massive" DB. These are:-

1. Thin sheet usually has a good surface finish as an "as received" material (R_A Circa 0.5 μm) and therefore requires limited pre-preparation. (This is particularly true of titanium alloy sheet).

2. Bonding loads can be applied by gas pressurisation which facilitates intimate contact between faces to be bonded and is therefore independant of the platen or workpiece flatness. Figure 35.

   (Mechanical bonding can also be applied for thin sheet bonding. The success of this process does however depend upon the platen and workpiece flatness (10)).

When associated with sheets of superplastic quality, conformity between faces to be bonded can be readily achieved by the fact that the material will form to provide intimate contact. In addition, the bonding pressures will remain normal to the surfaces of the sheets irrespective of the shape of the two sheets to be bonded.

5.9.3 SPF/DB

Since the late sixties, much experimental work has been conducted in the potential use of the combined process of SPF/DB, in particular, using the 6/4 titanium alloy. The structural forms which can be manufactured using these combined processes are universally recognised in terms of the numbers of sheets employed in their manufacture. Figures 36 to 38 illustrate the 2, 3 and 4 sheet forms respectively.

The manufacture of thin sheet SPF/DB components, in particular the two and three sheet forms rely upon selective DB prior to SPF. The achievement of the selective DB is associated with the use of an insoluable surface coating which is applied to the mating surfaces in areas which are not required to be bonded. In the case of the four sheet structure, the core line bonding can be carried out in a similar manner to that for the two and three sheet forms and prior to its enclosure into the skin pack. This structure also has a DB cycle as the completion of forming in the tool.

Although the full range of SPF/DB structural forms have already been demonstrated in association with titanium alloys, no published evidence exists to show that the three or four sheet structures can be applied to aluminium alloys. Because of the support provided by the tool to the bond areas employed in the two sheet structure, such a structure seems to be a practical proposition for SP aluminium alloys.

5.9.4 MMC

Because of the form in which these materials are produced (thin foil), coupled with the large areas of bond required, DB of MMC either as selective reinforcement in association with SPF/DB or in the manufacture of MMC components, generally uses gas pressure bonding to ensure good quality bonds. The use of these materials is still in the early development stage but the attractive weight saving potential provided by these high specific strength materials suggests that in the longer term this could represent the greatest potential use for DB in future aircraft component manufacture.

6.0 Component Integrity

Component integrity in respect to the use of DB in manufacture is dependent upon the type of structure and the method of manufacture employed in the production of the DB joints.

As discussed previously the types of structure and the high levels of joint loading associated with "massive" DB places considerable demands on the process conditions, tooling integrity and accuracy/finish of the component elements.

In general the DB joints associated with the thin sheet structure forms which are typical of SPF/DB manufacture, replace conventional joining technique such as welding, rivetting, or adhesive bonding. The load transfer demanded by such joints is well within the strength capability of DB joints, as demonstrated in section 5.6, where a comparison is provided between conventional and DB joint strengths.

Testing of practical SPF/DB structure has demonstrated however that the integrity of these thin sheet components is more dependent upon local stress concentrations in the parent material which arise from structural features which are typical and peculiar to this form of construction.

Figure 39 provides details of what are typical SPF/DB structural features. Although the use of DB has elimina' . the stress concentrations that would be associated with conventional jointing techniques such as rive.ting or bolting, it is clear from the sections presented in figure 39. that SPF/DB structures will be subject to stress concentrations. Figure 40 provides S-n data for one of these typical sections when subject to a typical loading mode. This data is compared with standard notched fatigue data and demonstrates quite clearly the presence of stress concentrations in the SPF/DB joint which are similar in value to a standard notch concentration. Similar data has now been derived for different loading modes including vibration loading and for the full range of structural features. This data provides the fundamental data base for the design of SPF/DB structures.

As indicated in tables 2 and 3, a number of full size component tests have been successfully carried out without adverse comment with respect to the integrity of the SPF/DB structures.

One particularly significant feature of the testing which has been conducted to date is the apparently excellent performance of SPF/DB structures in hostile acoustic environments. The range of components which have been tested are listed below and include both 2 and 4 sheet structures.

APU door for B1 - two sheet (33)

Nozzle panel for Concorde - two sheet (figure 41)

F-15 nozzle fairing - two sheet (7)

AV8B access door - four sheet (figure 42)

In all cases noise levels up to 165dB have been applied. In the case of the AV8B panel which is being developed by British Aerospace, this level was sustained for up to 40 hours without failure and in association with structural temperatures up to 150° C. These results clearly indicate a potential for the application of SPF/DB in the manufacture of cowling and nozzle structures or in fuselage areas which are subject to acoustic loading.

Service evidence of the integrity of SPF/DB structures is now being rapidly accumulated e.g. in the case of the Airbus wing access panels (figure 20) which were first fitted to the A310-300 aircraft, fleet time as of Spring 1987 was 26,000 hours with a lead aircraft flight time of 3200 hrs. Flight experience in support of the satisfactory integrity of SPF/DB structures is being accummulated on Tornado F-15, T-38, YF-12, F-4, F-14A, EAP, Rafale and RB211 engines.

Fatigue and damage tolerance has and is being addressed. In the case of the BLATS programme, it is reported (16) that two lifetimes of fatigue and two of damage tolerance were sustained without failure. Fatigue and damage tolerance testing has commenced on the primary fuselage pressure shell component (escape hatch) illustrated in figures 43 and 44. This component has been developed by British Aerospace and the testing will involve the application of some 120,000 cabin pressure cycles. Results of this testing will be available late 1987.

7.0 Design/Manufacturing Aspects Associated with SPF/DB Structures

All of the SPF/DB structural types which are presented schematically in Figures 36 and 38 have specific structural and manufacturing aspects which should be understood by the designer.

7.1 2-Sheet Structure

The two sheet structure of the type illustrated in figure 36 represents the SPF/DB equivalent of a common conventional form of construction and therefore, with the exception of the use of DB as a means of joining the two sheets, the structural advantages and limitations should already be well understood. Although the formed sheet will be subject to thinning, generally the shapes demanded in such structures do not present difficulties for the prediction of the thickness distributions.

In effecting the DB of such structures, two methods of manufacture are possible:-

a) pre bonding, either singly or in a pack, using gas pressure and delineating the bond areas using a bond-inhibitor (stop off)-then subsequently expanding in the form tool as a separate operation.

b) bonding by mechanical means in the tool prior to forming. This is usually a continuous but sequential operation.

The former route requires good alinement between the bonded areas of the blank and the corresponding features in the tool to ensure a consistent and optimum transition between the swaged section and the bonded area (reference figure 41).

Although the latter route will have a bond area which is defined and effected by the pressure generated by the tooling features, the quality of the bond is totally dependent upon the matching of tool faces, including the flatness and combined thicknesses of the component blank in the area to be bonded.

The designer must be aware therefore of the quality of bond and the resultant local features/defects that could result from the manufacturing route proposed. This is discussed in some detail in Reference (10).

7.2 3-Sheet Structure

Unlike the two sheet SPF/DB form, which is controlled directly by the tool in terms of its final structural definition, the three sheet structural form is largely controlled by the definition of the starting blank. The tool in this case, only provides a restraint to the blank around its periphery and in addition provides the envelope shape of the finished component. This therefore places responsibility on the designer to define both the starting blank and the finished component and to accomplish this, it is necessary for the designer to model and understand the forming process. For simple structures, such as that illustrated in figure 38, the modelling is relatively easy but for the more complex stiffening patterns such as sine wave or discontinuous stiffening, the establishment of a finite element technique as a basis of an SPF CAD system i highly desirable and will greatly reduce the iterative trial and error development which is currently a feature of SPF/DB technology.

The fact that the tooling can, to a large degree, be divorced from the component structural definition, provides the designer in turn with freedom to change the structural definition without affecting the tool. This is comforting to the designer bearing in mind the cost and lead time of provisioning tools.

One feature of the three sheet structure, which needs to be observed by the designer in the definition of the blank, is the fact that the skin thicknesses used need to be significantly greater than the core thickness to ensure skin stability during forming and to overcome external waviness on the finished component.

A wide variety of core stiffening patterns are possible with the three sheet structure. These include straight, sine wave, and discontinuous features. Orthogonal stiffening is possible for a three sheet structure but requires considerable increase in the blank complexity.

7.3 4-Sheet Structure

The four sheet structure illustrated in figure 37 results in cellular stiffening which has the virtue of having design and stressing characteristics which, although not identical in detail, are familiar to the designer in their conventionally fabricated equivalent.

As for the three sheet structure, the designer has the responsibility of defining the details of the starting blank to achieve his finished component. Again this cannot be achieved without an in depth understanding of the manufacturing process and will be greatly assisted by the development of a finite element CAD analysis system.

In the Authors view the four sheet structural form provides the widest range of freedoms to the design in the derivation of his optimum design. These are:-

- the ability to change the skin envelope thicknesses and associated reinforcement independent of the core-including the deletion of either or both skins to produce two or three sheet cellular forms.
- the ability to change the core thicknesses and associated reinforcements independent of the skin.
- the ability to change the number, form and position of stiffeners.
- the ability to interrupt stiffeners.
- the ability to have orthogonal stiffening patterns.
- the ability to form integral fittings ref. figure 45.
- the ability to facilitate access for control runs etc. by forming web apertures concurrent with the forming of the component ref. figure 46.

8.0 NDT Aspects of SPF/DB Structures

There are essentially three aspects to the NDT of SPF/DB structures. These are

a) thickness measurement

b) inspection of internal details

c) inspection of DB

For all of the SPF/DB forms, ultrasonic thickness measurement has proven to be reliable in production. This cannot however facilitate all of the measurements, in particular for the internal structure in the three and four sheet forms. These can only be inspected by component cut up on a sampling basis, coupled also with close control of the process parameters and particularly the accuracy of the starting blank. SP quality of the material is also an important aspect of maintaining control of the finished component thicknesses.

Inspection of the internal details of a formed structure, in particular the three and four sheet forms, can be carried out accurately by X-ray. With close control of the blank and its processing, coupled also with thickness measurements and external witnesses, it is likely that such inspections may be reduced to a sampling basis.

Ultrasonic C-Scan represents the only practical production NDT technique available for the inspection of DB quality. This method can be readily applied to the two and three sheet forms and to the skin to core bonding of the four sheet structure. Bond quality of the internal structure of the four sheet form can only be checked by component cut up on a sampling basis, coupled also with an inferred quality from the results of the inspection of the core to skin bond which is produced in identical processing conditions and environment to the web bonds.

9.0 Economic Advantages

The economic advantages associated with DB and SPF/DB have been well documented and have been substantiated fully by a wide range of full size development/demonstrator projects conducted by most aerospace companies throughout the Western World.

The fundamental aspects associated with these economic advantaged are given in Tables 6 and 7 but the essential elements are -

1. Simple starting blank forms (particularly significant for titanium).
2. High material utilisation
3. Reduced parts count.
4. Process times which are insensitive to size, complexity of structural form, or numbers of components manufactured in one operation.
5. In the case of SPF/DB concurrent forming and bonding.

It is recognised that the actual savings achieved are a function of the component, its equivalent conventional method of manufacture and the associated material. A highly significant factor in the maximisation of these savings is however a requirement to design for these processes - in particular SPF/DB - rather than slavishly attempting to reproduce the conventional equivalent.

As an example of the significant reduction in parts count which is attainable by SPF/DB, the pressure shell escape hatch illustrated in Figure 44 and which has been development by British Aerospace for 125/800 aircraft, has a reduction of 76 detail parts and 1,000 fasteners in transforming from aluminium fabrication to titanium SPF/DB. The SPF/DB door will have only 14 details and some 90 fasteners. Allowing for differences in the raw material costs between the two methods of manufacture a 30% cost saving is predicted for the SPF/DB door. This component is a further example of the competitiveness of SPF/DB titanium in the replacement of aluminium fabricated structure. Further examples of the types of components which have been replaced by SPF/DB are given in tables 2 and 3.

10.0 Weight Advantage

Again the weight saving potential provided by these processes has been well documented and substantiated by full scale component manufacture. These weight savings occur from the ability of SPF/DB in particular, to produce efficient structural forms with the elimination of fasteners and associated joint flanges. A particular example is shown on Figure 20. This shows the titanium SPF/DB wing access panel which has been developed by British Aerospace and is now in full production for fitment to Airbus A310 and A320 aircraft. This particular component achieves a weight saving of 40% of the aluminium alloy equivalent.

11.0 Applicability

Based upon an assessment of civil and military aircraft, an estimate of the potential embodiment levels for components using the SPF/DB process route is seen as follows:

	% of Structure Weight
SPF/DB	8 - 10

Almost half of this application is seen in the replacement of aluminium fabricated structure. Areas of application are:-

SPF/DB
- control surfaces
- smaller flying surfaces
- access panels/doors
- engine bay component
- hot ducting
- engine rotating parts

12.0 Conclusions

SPF and SPF/DB of structural alloys, in particular titanium, have come of age, in as much as these processes have now been introduced into the volume production of a significant number of primary components for airframes and aero engines.

The processes are well understood and produce components which are superior in strength, cost and quality, to their conventionally manufactured equivalents.

Conventional NDT techniques can and are already being successfully applied in the qualification of production components but further R&D in the field of detection of intimate contact disbonds is desirable. In the final analysis the key to quality of SPF/DB structures is seen as close process control.

To exploit these processes, an indepth understanding of their flexibilities and limitations is required by the designer if the maximum benefits are to be realised. The designers capabilities will be greatly enhanced by the successful development of a CAD/CAM modelling system for these processes.

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TABLE 1

Aircraft Superplastic Alloys

Aluminium	Titanium
Superal 100	6 Al-4V (IMI 318)
Superal 150	4 Al-4Mo -2.Sn-0.5Si (IMI 550)
Superal 220	CORONA 5
7475	6Al-2Sn-4Zr-2Mo
Al-Lith	15V-3Cr-3Sn-3Al
Nickel	8Al-1Mo-1V
INCO 718	3Al-2.5V
IN 100	5Al-2.5V
RSR 143 and 185	

TABLE 2

SPF & SPF/DB DEVELOPMENT COMPONENTS

COMPONENT/AIRCRAFT	FORM (Titanium)	REPLACING	COST SAVING %	WEIGHT SAVING %	TEST CONDITIONS	COMMENT	REF.
BLATS - Wing Carry Through structure	SPF/DB Assembly	Titanium Fabrication	41	30	2 lives fatigue 2 lives damage Tolerance	Experimental Design	(7) (16)
Nacelle Frames - B1	SPF SPF/DB	Titanium Fabrication	55 43.5	33 40		Integrally stiffened webs	(7) (16)
APU Door - B1	SPF/DB	Titanium Machining	50	31	Acoustic Test duration 15 hrs levels 159-167dB	In production B1-B but SPF only	(7) (116) (33)
Engine Bay - B1 Door	SPF/DB	Aluminium Honeycombe	34	29		Largest component made to date 105" x 60"	(7) (16) (33)
Hot Air - B1 windscreen blower	SPF/DB	Steel Fabrication	40	50	Fully qualified for B-1	In production for B-1B	(7) (16) (33)
Undercarriage - T38 Door	SPF/DB	Aluminium Honeycombe	2	$5m life cycle costs	Fully qualified for flight and flight tested since early 1980's.	Likely full scale production as replacements	(6) (7)
Horizontal - F4 Stabiliser T.E.	SPF/DB	Steel Honeycombe				Proposed for Flight Test	(7)

Table 3

SPF & SPF/DB DEVELOPMENT COMPONENTS

COMPONENT/AIRCRAFT	FORM (TITANIUM)	REPLACING	COST SAVING %	WEIGHT SAVING %	TEST CONDITIONS	COMMENT	REF.
T.E. Flap - C-17	SPF/DB		15	26	Fully qualified for flight.		(6) (7)
Nozzle Fairing - F-15	SPF/DB	Titanium Honeycombe	40		Acoustic testing Levels up to 165dB. In excess of three years flight testing.	1st SPF/DB component to be flight tested. Features in redesign of F-15 nacelle area.	(6) (7)
Wing Glove Fairing - F14A	SPF/DB	Aluminium Fabrication	30	10		Proposed as flight demonstrator	(6) (7)
Escape Hatch - BAe 125/800	SPF/DB	Aluminium Fabrication	30	10	Fatigue test to 120,000 pressure cycles.	Ref. figure 44	
Access Panel - Harrier AV8B	SPF/DB	Aluminium Fabrication plus titanium skin		12	Acoustic Testing Duration 40 hours Levels - 165dB	Ref. figure 42	
Slat Surface - Rafale	SPF/DB	Aluminium Fabrication			Fully qualified for flight test.	Experimental	
Hot Ducting - Tornado	SPF/DB	Titanium Fabrication	34				(9)
Missile Wings -	SPF/DB	Various					(8)

TABLE 4

ADVANTAGES OF SPF

AND SPF/DB

- WEIGHT SAVINGS UP TO 50%
- COST SAVINGS UP TO 60%

THE ACTUAL SAVINGS ARE A FUNCTION OF

- COMPONENT
- ALTERNATIVE METHOD OF MANUFACTURE
- MATERIAL USED IN CONVENTIONAL MANUFACTURE

TABLE 5

ADVANTAGES OF SUPERPLASTIC FORMING (SPF)

Characteristic	Cost benefits	Structural benefits
Forming capability in excess of conventional forming	One-piece forming rather than multipart joining	Reduction in joints for compression structure low cost stabilizing features
'One shot' process	Reduction in tooling and/or intermediate heat treatment	—
Accuracy/repeatability	Reduction in assembly/fitting costs	Low variability in structural performance
Simple cavity tooling	No requirement for matched tooling	—
Negligible tool wear	Reduced tool refurbishment/replacement	—
Process time independent of size or number	Increasing benefit with size or use of multicavity tooling	—

TABLE 6

ADVANTAGES OF DIFFUSION BONDING (DB)

Characteristics	Cost benefits	Structural benefits
Process time independent of bond area or numbers	Increasing benefits with sizes or multiples off in one press cycle	—
Joint strengths approaching or equal to parent metal	—	Improved joint efficiency relative to conventional joining
Low strain/distortion	—	—

TABLE 7

ADVANTAGES OF SPF/DB

Characteristics	Cost benefits	Structural benefits
Simple starting blanks	Of particular importance for titanium	—
Concurrent SPF/DB processes	Joining and forming as combined operation	—
Complex monolithic structures	Reduction in details and fasteners	Reduction in joint lands
Combines benefits of SPF and DB as independent processes	Ref. advantages of the separate processes	Ref. advantages of the separate processes
High material utilization	Of particular importance for titanium	—

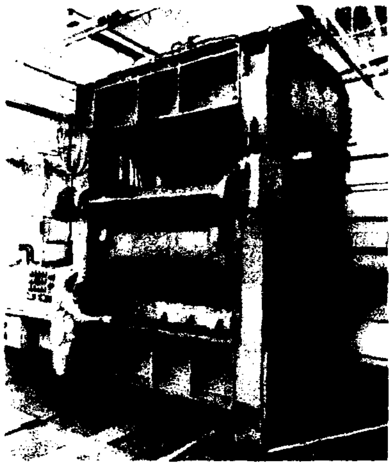


Fig1

HEATED PLATEN PRESS

Male Forming of Aluminium
(T.I. Superform Method)

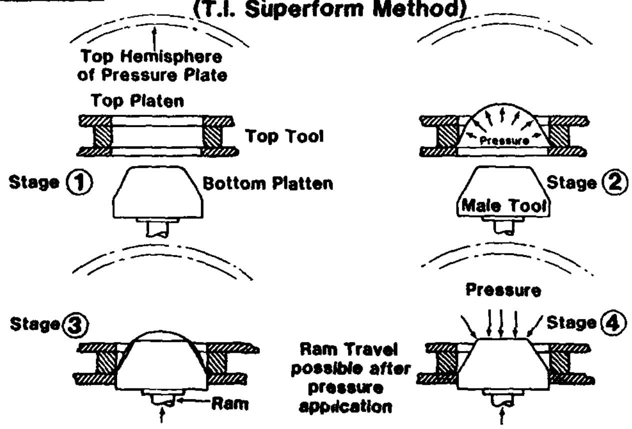
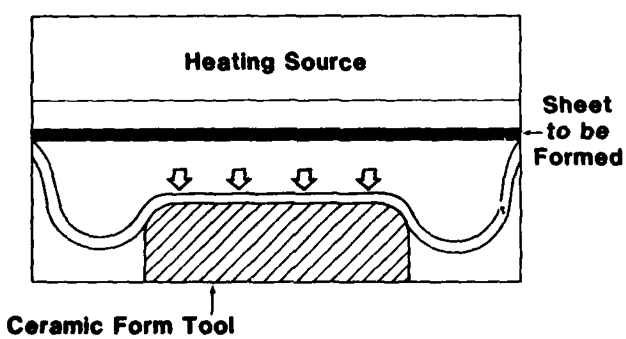


Fig2

"Hot Box" Forming

Fig3



Membrane Forming

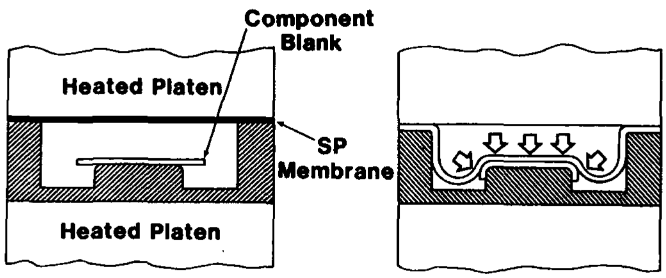
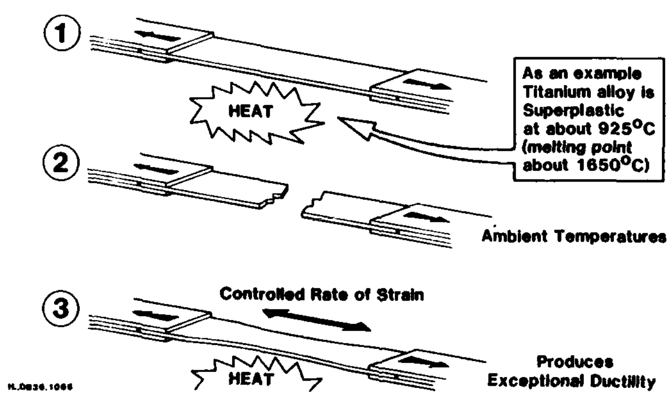


Fig 4

Fig5 - Refer Figure 20

Fig6

The Basic Processes of 'Superplasticity'



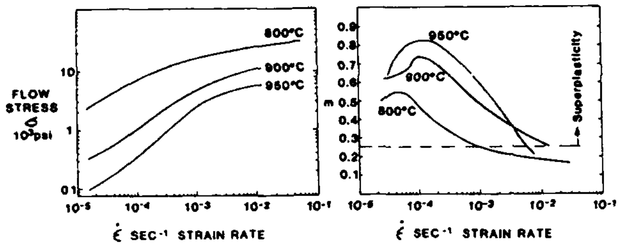
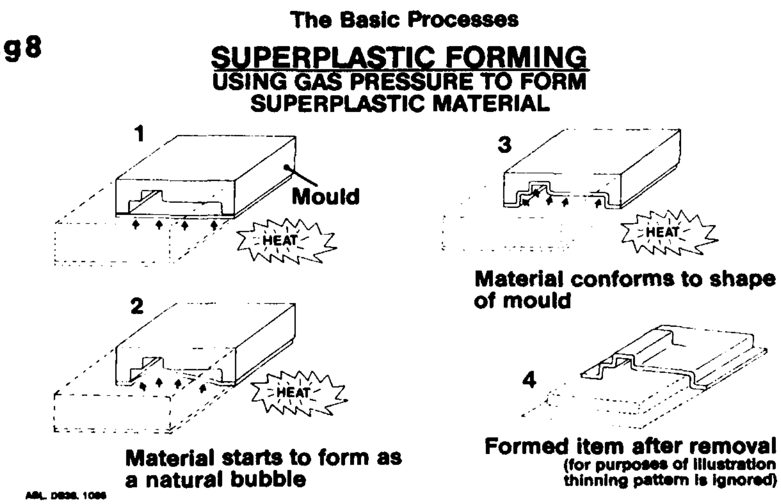


Fig7 Superplastic Material Characteristics

(Titanium 6/4)

Fig8



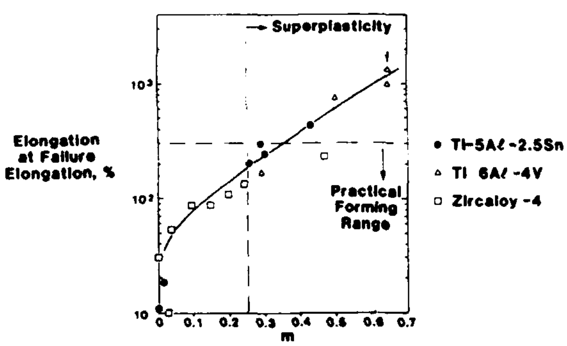


Fig9 Elongation to Failure of Superplastic Materials

WIDE COLUMN CONCEPTS

Fig 10

The Effect of Various Structural Configurations on Compressive Efficiency, ϵ

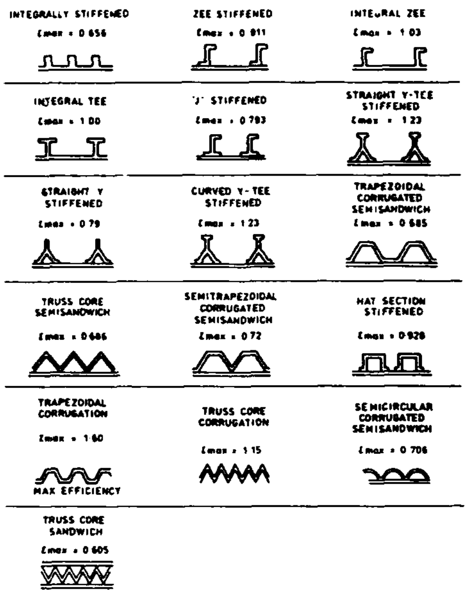
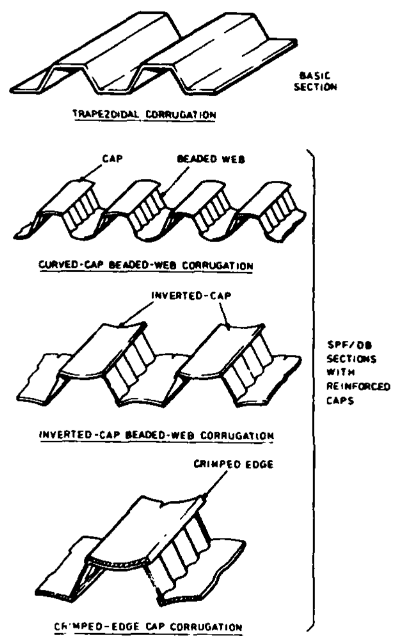
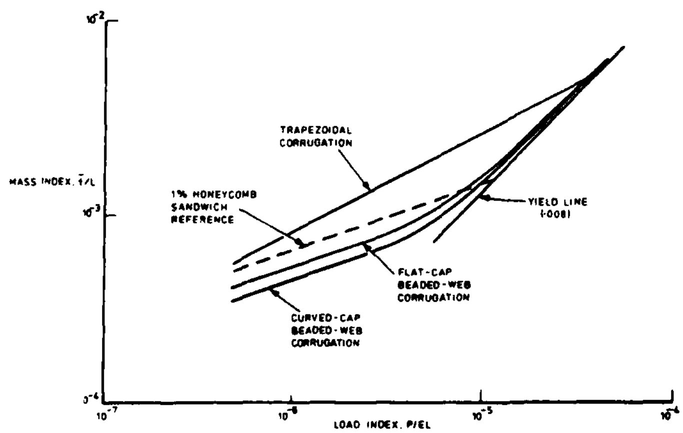


Fig 11

THE USE OF SUPERPLASTIC FORMING TECHNOLOGY TO IMPROVE THE BASIC TRAPEZOIDAL CORRUGATION





EFFICIENCY COMPARISON FOR CURVED-CAP BEADED-WEB CORRUGATION AND FLAT-CAP AND TRAPEZOIDAL CORRUGATIONS

Fig 12

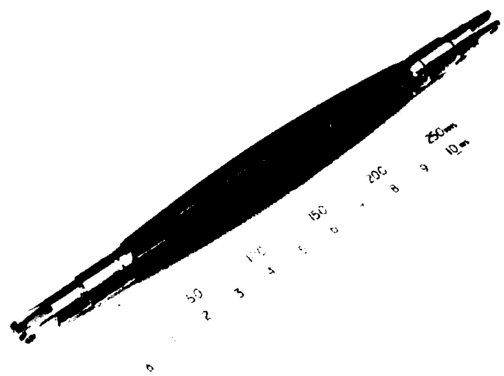
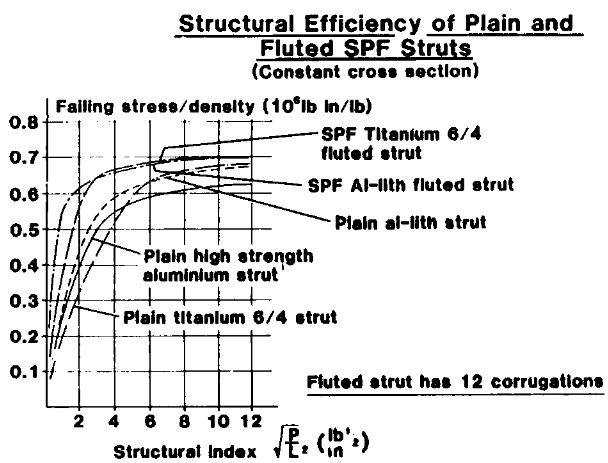


Fig 13

SPF FLUTED STRUT

Fig 14



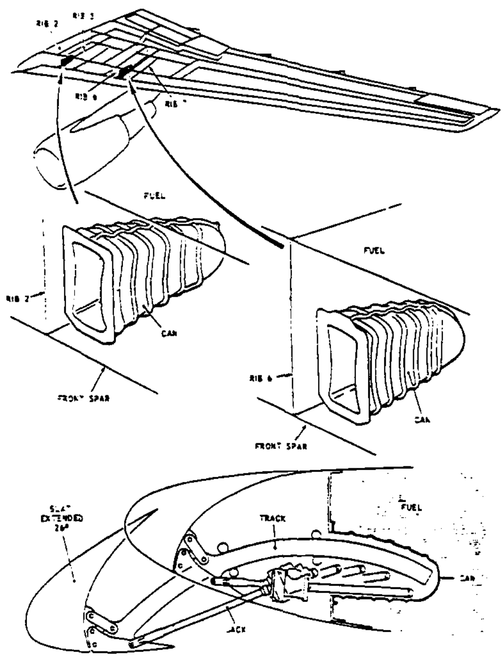


Fig 15

Location of Airbus A310 Jack ~an

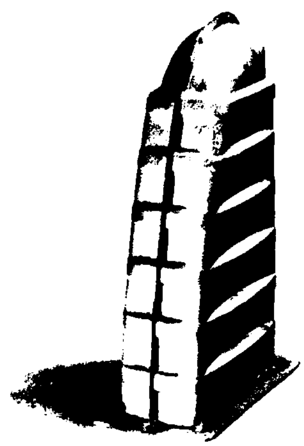


Fig 16

SPF Jack Can (Titanium 6/4)

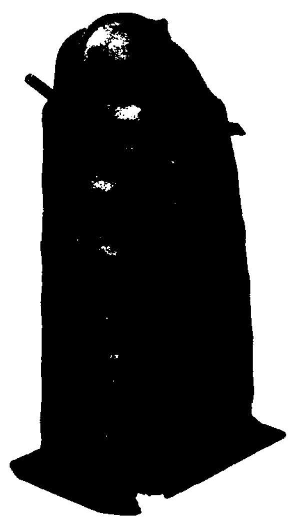


Fig 17

Conventionally Pressed Jack Can
(C.P. Titanium)

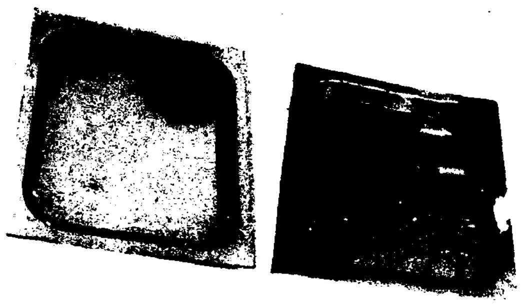
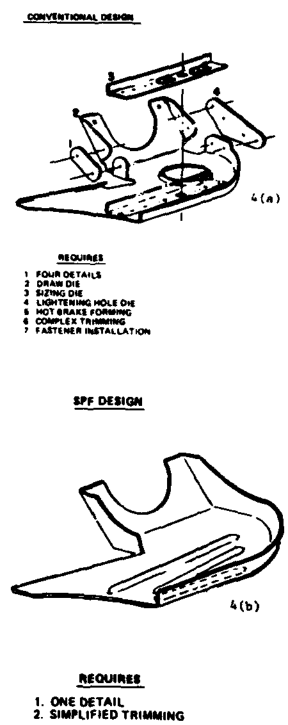


Fig18

Concorde Engine Bay Panel

Fig19

Ejector Valve Bracket
(Northrop)



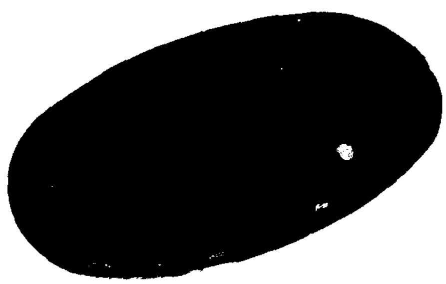


Fig20 Airbus A310/A320 Wing Access Panel

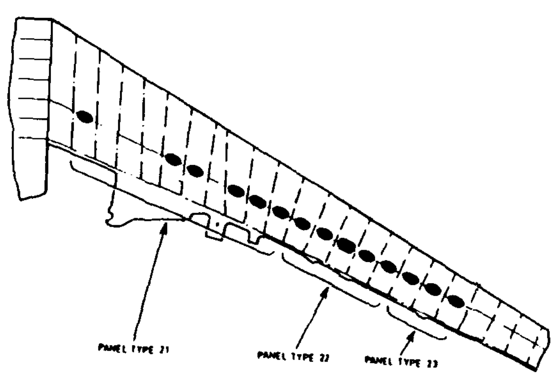


Fig21 Location of Airbus Wing Access P

Superplastic Forming Thickness and Time Penalties Associated with Fine Detail

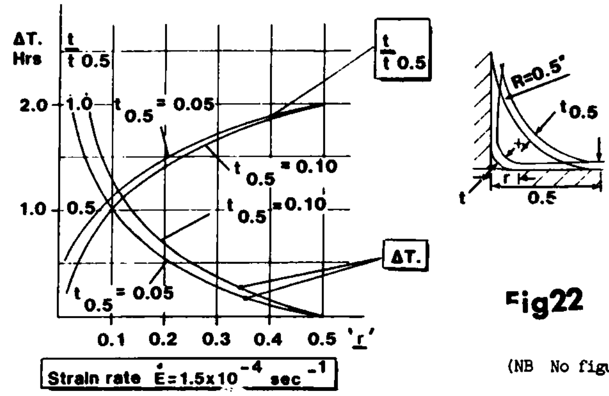
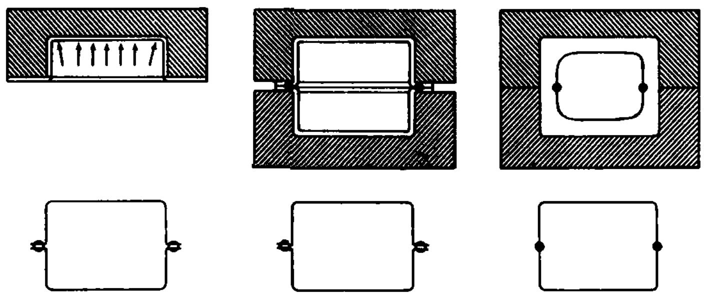


Fig22

(NB No figure 23)

Use of Simple Cavity Tools

Fig24



Manufacture of Two Pressings assembled and joined

Forming of Complete Envelope joined at the blank stage

Use of a Preform to reduce thinning

Forming in Relation to the Material SPF Characteristics

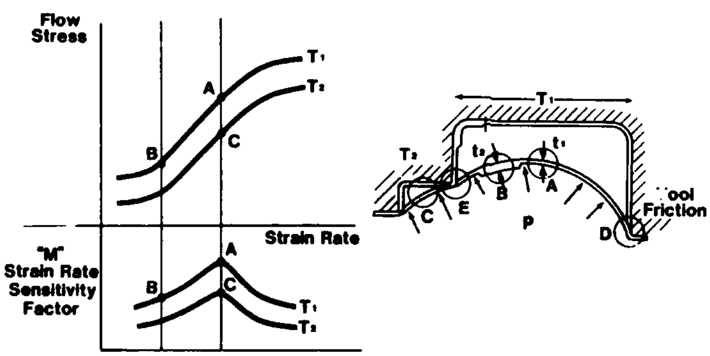


Fig25

Male Forming/Female Forming

Fig26

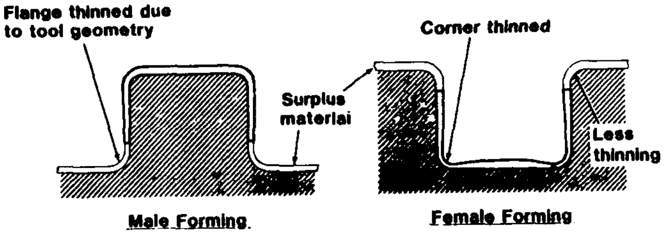
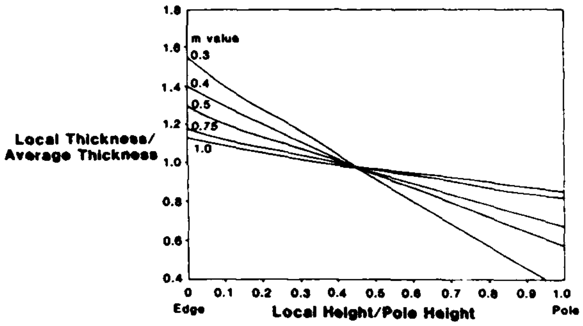


Fig27

Effect of SP conditions on Thickness



Process	Roughness R_a in µm							
	0.05	0.1	0.2	0.4	0.8	1.6	3.3	6.3
Lapping	▬	▬	▬	▬				
Polishing		▬	▬	▬				
Grinding		▬	▬	▬	▬	▬		
Turning				▬	▬	▬	▬	▬
Milling					▬	▬	▬	▬

Typical Roughness Values
Obtainable by Machining Processes

Fig28

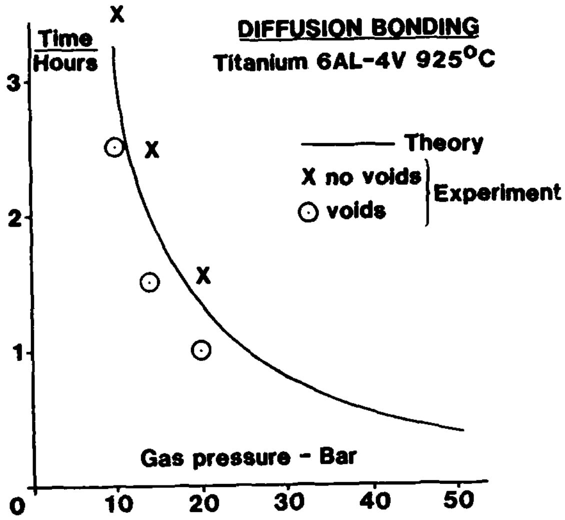
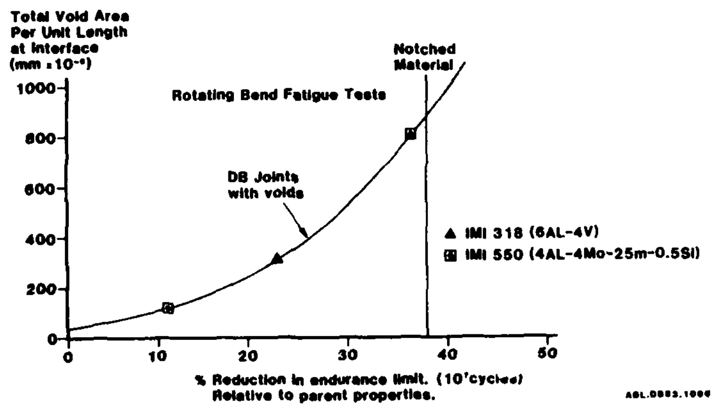


Fig29

Diffusion Bonding-Effect of Micro Voids on Fatigue
(Ref. Partridge & Baker)

Fig30



Diffusion Bonding - Effect of Micro Voids on Impact Strength
(Ref. Partridge & Baker)

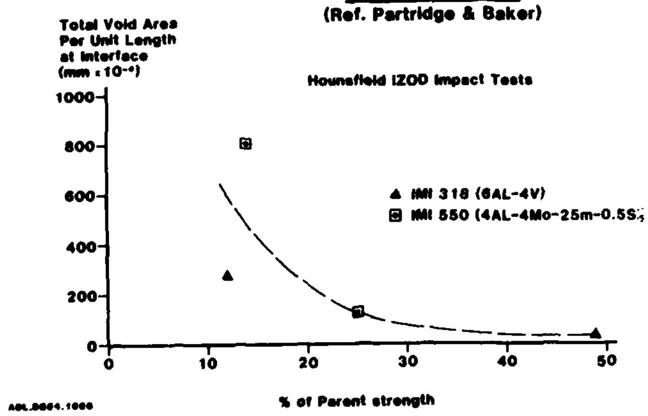
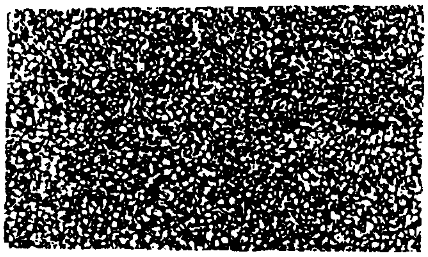


Fig31

Diffusion Bond Defect

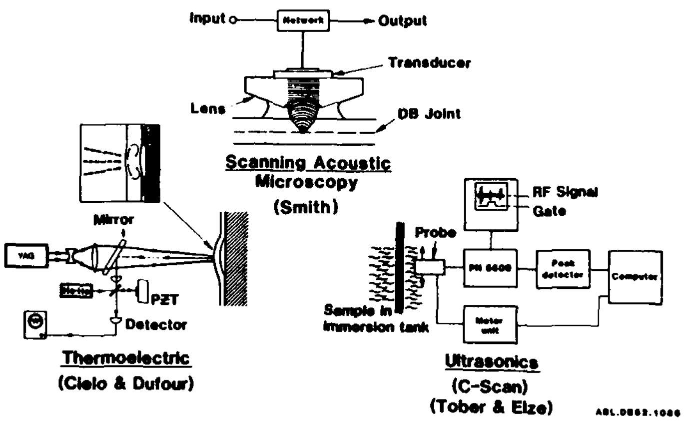
-Micro voids

Fig32



Diffusion Bonding - N.D.T. Techniques

Fig33



Massive DB **Massive DB**

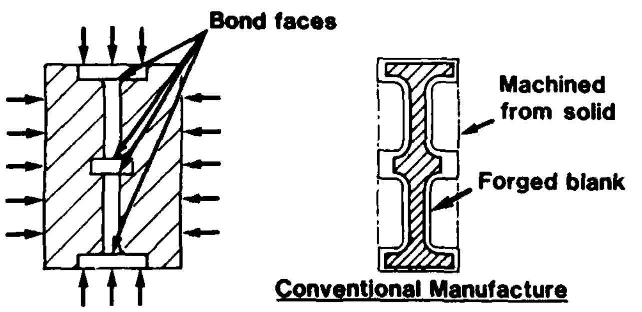


Fig34

Thin Sheet DB-Gas Pressure Bonding

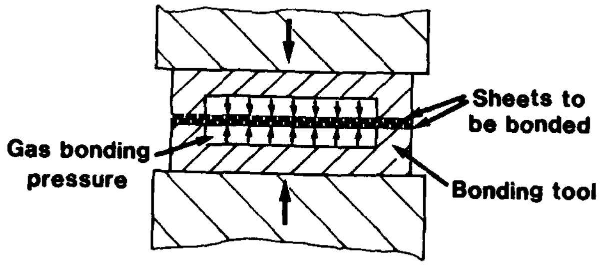


Fig35

Fig37 Four Sheet SPF/DB Structure

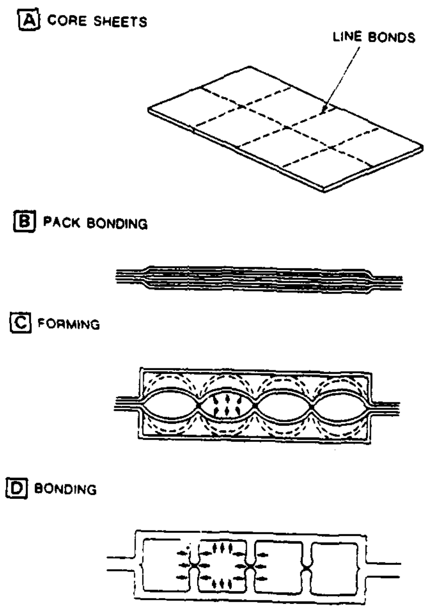


Fig36 Two Sheet SPF/DB Structure

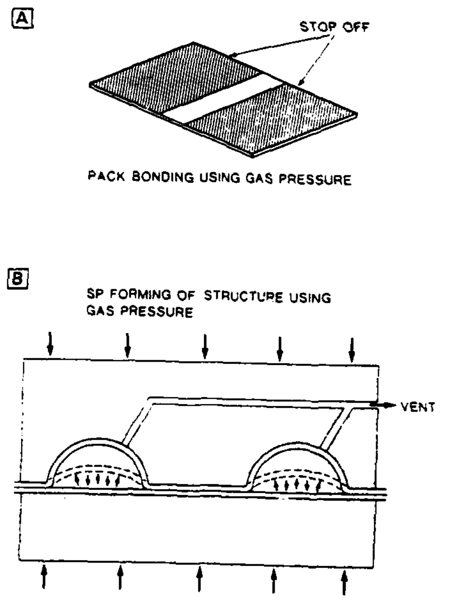
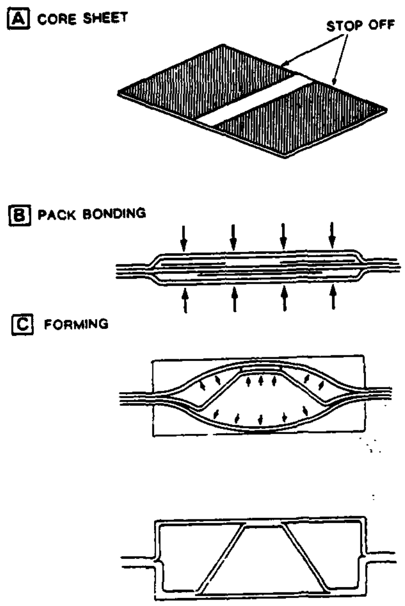
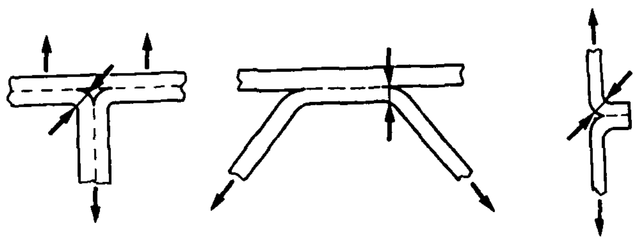


Fig38 Three Sheet SPF/DB



SPF/DB Structural Details **Fig39**

⟷ **Typical failure locations**



Fatigue Strength of Typical SPF/DB Joint **Fig40**

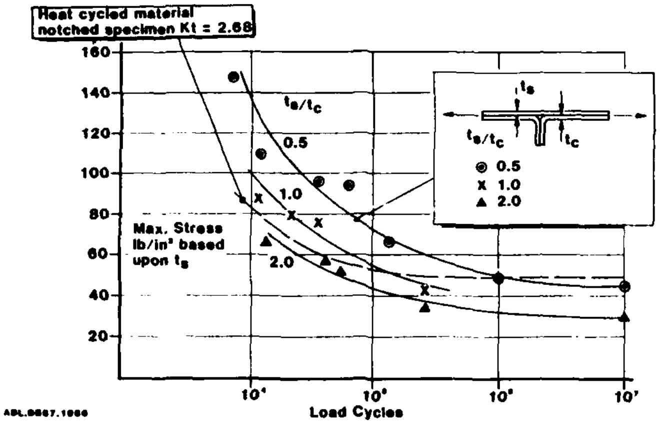
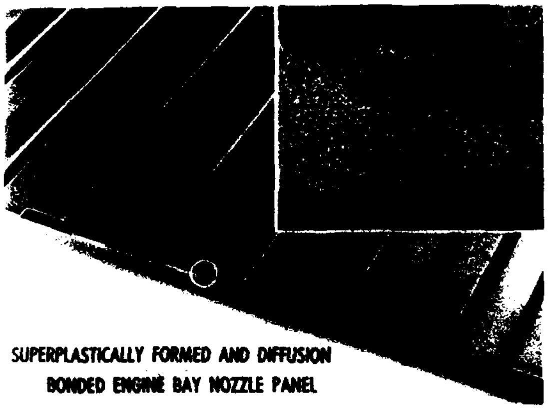


Fig41



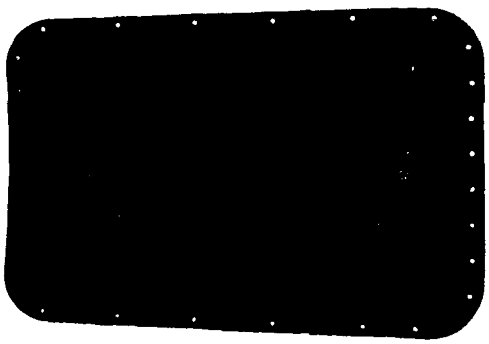


Fig42

AV8B/Harrier
SPF/DB Access Door

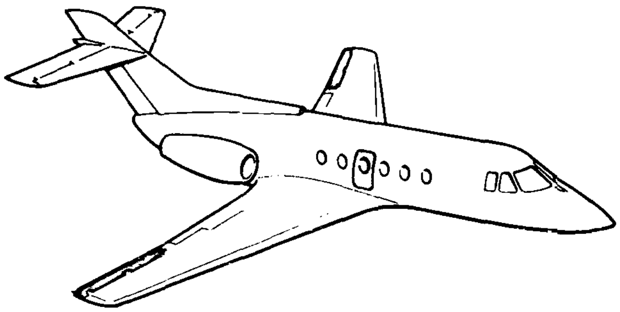


Fig 43

British Aerospace
125/800
(Location of Escape Hatch)

SPF/DB Escape Hatch

Fig 44

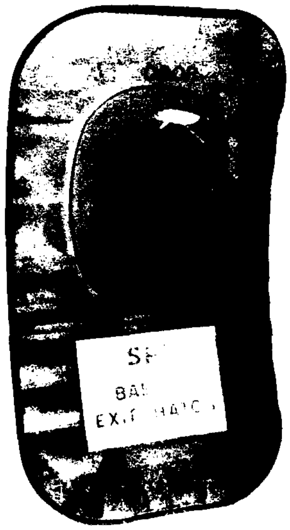


Fig 45

Four Sheet SPF/DB
(Integral Fitting)

Fig 46

Four Sheet SPF/DB
(Web Apertures)

THE MANUFACTURE OF SUPERPLASTIC ALLOYS

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BANBURY Oxon OX16 7SP
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SUMMARY

The requirements in a superplastic sheet material are either to possess a fine uniform grain size or to be *capable of developing* such a grain structure during the course of superplastic deformation. Alloys that are currently being commercially exploited fall into both categories and the manufacturing routes to produce material with good superplastic behaviour using commercial scale processing are considered. Most of the paper is devoted to aluminium alloys since their manufacture in a superplastic form embodies most of the general principles. Since the aluminium alloys do not involve allotropic transformations limited consideration is also given to titanium alloys. Amongst aluminium alloys, the Al-6% Cu-0.4%Zr system that typifies the Supral alloys, requires a specially developed casting system in order to achieve a very high level of supersaturation with zirconium. Subsequent manufacturing is fairly conventional, the sheet product recrystallising *dynamically during superplastic forming.* The higher strength aluminium alloys are conventionally cast but achieve a very fine grain size at the sheet stage by careful thermal mechanical-treatment during the later stages of semi-fabrication. With titanium the standard production route for the Ti-6Al-4' sults in a product with superplastic capabilities perfectly adequate for most applications.

INTRODUCTION

While the phenomenon of superplasticity has been recognised for at least sixty years it has only been the subject of intensive investigation since the appearance of Underwood's review of the phenomenon[1] in 1962. During the intervening twenty-five, or so, years the literature has multiplied greatly and, as pointed out by Nieh and Wadsworth[2], this growth has been accompanied by a change of emphasis from "academic" alloys to "commercial" alloys. Despite this change of emphasis, the overwhelming majority of the published literature has been concerned with mechanisms of superplastic deformation while most of the remainder has been concerned with superplastic forming. This, of course, comes about because of the commercial sensitivity of information relating to the more practical aspects of superplasticity and, for the same reasons, very little has been published relating to the techniques of manufacture of superplastically deformable metallic materials. In this paper an attempt will be made to give, at least some, description of the aspects that require control from the point of view of the metal manufacturer. Table 1 indicates some of the alloy systems that had been shown to be superplastic by 1970 while Table 2 includes some additions for the intervening years. In a paper of this type it is clearly impossible to consider all the possible materials and most attention will be devoted to the two most important sets of alloys, namely those based on titanium and those based on aluminium. Most detail will be given on aluminium alloys as the various superplastic aluminium alloys probably incorporate most of the manufacturing principles of importance in superplastic alloy systems, regardless of base metal. Superplastic titanium alloys then embrace the few remaining principles of importance.

REQUIREMENTS FOR SUPERPLASTIC BEHAVIOUR

The early view of superplastic alloys was that they should consist of two phases present in approximately equal volumes and the two phases should be very finely divided. Thus, it was felt, the alloys should be of eutectic or eutectoid composition. Although, even from an early stage, exceptions to this general rule were noted (e.g. superplastic behaviour was observed in commercial purity zinc in 1933[3]), it was probably the demonstration of good superplastic behaviour in Al-Cu-Zr alloys [4] that led to greatest modification of the early beliefs. The Al-Cu-Zr alloys were, at the superplastic forming temperature, virtually single phase but they contained a dispersion of very fine (~ 10 nm) particles of Al_3Zr to confer grain stability. This observation led, in due course, to the demonstration of superplastic behaviour in several conventional aluminium alloys after *special thermo-mechanical* processing. Overall it is now clear that to be capable of superplastic behaviour <u>any</u> alloy must:

- either possess a stable, fine grain structure before deformation commences
- or be capable of achieving a stable, fine grain structure during the course of plastic/superplastic deformation.

Additionally, the flow stress during superplastic deformation is of considerable importance, particularly in those alloys with a significant cavitational tendency. This is because in forming structural components it becomes essential to superimpose a hydrostatic stress in order to suppress formation of cavities and, it has been suggested, a superimposed hydrostatic stress approximately equal to half the flow stress of the material is required. Thus the necessary hydrostatic stress to suppress cavitation will go up as the flow stress of the alloy goes up and this can have profound implications for the forming machinery. In the manufacture of structural aerospace components a forming machine with a plan area of 1½ metres x 3 metres could well be required. For a forming operation, without superimposed back pressure, this would imply a maximum separating force of the order of 400 tonnes. For a superplastic aluminium alloy with a flow stress of 5 MPa the superimposed hydrostatic stress to suppress cavitation would be 2.5 MPa which implies a separating force on the example machine of the order of 1130 tonnes.

It will, almost invariably, be the case that as strain rate goes up so will flow stress and the natural wish of the component manufacturer to carry out the forming operation as rapidly as possible may well bring with it larger flow stresses with their inevitable consequences for forming machine design. Figure 1 (after Hamilton et al[5]) indicates the extent of change in flow stress that may accompany strain rate changes. More than an order of magnitude change in flow stress accompanies an increase in strain rate from the optimum for superplastic behaviour (4×10^{-4} sec^{-1} for the example alloy) to that at which it might well be wished to carry out the forming operation (say, 1×10^{-2} sec^{-1}). The separating force on the machine in the earlier example now goes up to well over 2000 tonnes if cavitation is to be suppressed.

Temperature of forming also exerts a profound influence upon flow stress and this is also illustrated for 7475 aluminium in Figure 1. This, of course, has a two-fold effect for the superplastic forming machine. On the one hand, the lower the forming temperature the cheaper the materials from which the machine may be constructed. At forming temperatures of the order of 480°C relatively cheap cost aluminium tooling can be used and the constructional materials for the rest of the machine do not have to be exotic. As temperatures approach 600°C more expensive materials are necessary both for general construction and tooling. On the other hand, however, the higher the forming temperature the lower will be the flow stress and hence the requirements for superimposed back pressure will be less.

In summary, therefore, the objectives that the manufacturer of superplastic metal sheet should achieve are a material either possessing, or capable of developing during forming, a fine stable grain size with as low a flow stress as possible. The customer wishing to form the sheet will demand consistency of superplastic behaviour in terms of optimum ductility, forming temperature, strain rate and flow stress and the customer will also expect to find this consistency across a wide range of sheet gauges and a wide range of sheet sizes.

SUPERPLASTIC ALLOY MANUFACTURE

ALUMINIUM ALLOYS

The earliest examples of superplasticity in aluminium alloys concerned eutectics, such as Al-33% Cu, possessing unsatisfactory service properties and it was probably not until the invention of the so-called Supral alloys[4] within the Tube Investments/British Aluminium Group that the possibilities for superplastically deformable aluminium alloys became generally appreciated. Subsequently, work on aluminium alloys of standard composition, and notably that carried out at Rockwell International[6], demonstrated that by suitable thermal mechanical treatment, alloys such as AA7475 could possess good superplastic properties. More recently, great interest has developed in both the aluminium industry and the aerospace industry in alloys based upon the aluminium-lithium system and several of the new candidate alloys have been shown to be capable of superplastic behaviour. In the following sections the manufacture of these three groups of aluminium alloys will be considered and brief mention will be made of eutectic alloys. Somewhat different principles are involved between the first two groups while, it would appear, that the aluminium-lithium based alloys can, by appropriate processing, be placed in either of the first two groups. The alloys will be considered here in the same order in which they were developed since this should present the necessary information in a coherent manner. Figure 2 schematically illustrates the manufacturing route for conventional structural aluminium alloys so that in the subsequent sections comparisons can be made between the conventional route and the modifications necessary to confer superplasticity in different alloys.

The Supral Alloys

The trade name Supral was registered to describe the superplastically deformable aluminium alloys (7 8 9) manufactured by the former British Aluminium Company (now British Alcan Aluminium plc) and formed into engineering components by the Alcan subsidiary, Superform Metals Limited. Essentially, these alloys are all characterised by a zirconium content (approximately 0.4%) that is considerably higher than that found in conventionally manufactured aluminium alloys. Additionally, the alloys all contain a further addition (or additions) such as copper, zinc or magnesium. It has been postulated that the mechanism by which superplastic deformation is achieved in these alloys is for the major alloying addition, such as copper, to lower the stacking fault energy of the alloy, so encouraging dynamic recrystallisation, while the zirconium provides a dispersion of very fine particles that stabilise the newly recrystallised grains. The manufacturing route was developed around this hypothesis. The greatest problem to be overcome in making alloys of this type is the need to suppress the separation of coarse particles of Al_3Zr during casting. Figure 3 shows the binary aluminium-zirconium phase diagram and it can be seen that at a zirconium content of approximately 0.4% by weight, and under equilibrium conditions, the first phase to separate on solidification is Al_3Zr. Under such conditions the Al_3Zr could separate in a very coarse form commencing at a temperature of approximately 780°C and continuing until the peritectic temperature of 660°C. If solidification occurred under equilibrium conditions the structure of an Al-0.4%Zr alloy would consist of very large Al_3Zr particles in an aluminium-matrix containing a very low level of zirconium in solid solution in the matrix. The Al-Cu eutectic temperature is 548°C and this, therefore, defines the maximum temperature to which the solid alloy may be subjected if incipient melting is to be avoided. It is not possible, even by prolonged heating at very high temperatures, to achieve a solid solution level of zirconium any greater than 0.1%. Thus, to achieve the required degree of supersaturation with zirconium, solidification must take place under non- equilibrium conditions and sufficiently rapidly to prevent, or at least minimise, separation during solidification.

Figure 2(a) shows, diagramatically, a cross-section of the arrangement for the vertical semi-continuous, direct chill (DC) casting of aluminium alloys. Basically, the same arrangement is employed for the overwhelming majority of aluminium cast for fabrication into wrought products. An open bottomed water cooled mould, 1, is employed. The initial chill of the metal entering via the distributor, 2, is provided by contact with the mould wall but the great majority of cooling is achieved by the direct impingement of jets of water from the bottom of the mould onto the side of the partially solidified ingot, 3. In a normal casting operation the molten metal would enter the mould at a temperature of about 700°C eventually solidifying at the interface between the molten metal sump and the emerging ingot when the temperature will have fallen to about 660°C for pure aluminium or appreciably lower for many alloys.

In casting alloys with a zirconium content of the order of 0.4% the temperature of the metal entering the casting head must be much higher than the example quoted above since Al_3Zr is potentially, capable of separating once the metal temperature falls beneath about 780°C. Thus the molten alloy should enter the casting head at a temperature somewhat greater than 780°C but, once in the molten metal sump, must be chilled at as high a rate as possible in order to suppress the separation of the Al_3Zr. This, of course implies that the solidification of the ingot should take place as rapidly as possible which, in turn, increases the level of solidification stresses in the ingot and raises the probability of ingot cracking. To make fabrication ingot of the Supral alloys, therefore, a specially designed casting system has to be employed which minimises the residence time of molten metal in the sump, achieves solidification as rapidly as possible in a DC system but which does not result in high residual stresses in the cast ingot. Even employing such a system separation of coarse Al_3Zr is not wholly suppressed and Figure 4 illustrates the type of coarse Al_3Zr precipitation that can occur. Note that the secondary dendrite arm spacing is also very fine indicating the rapid solidification rate of the alloy.

After casting the next stage in the manufacturing route is to subject the cast ingot to a, so-called, homogenisation process. In this, conventionally, the ingot is held at a sufficiently high temperature to allow diffusion to take place and remove the compositional variations that exist across cast grains while coarse intermetallic compounds that separate during solidification are also dissolved. In the Supral alloys the high level of zirconium supersaturation in the casting, together with the low zirconium solid solubility means that any elevated temperature treatment results in precipitation of zirconium from the supersaturated solid solution. A conventional, high temperature homogenisation treatment would result in the formation of a relatively coarse dispersion of Al_3Zr particles that would not be very effective in controlling grain growth during superplastic deformation. The heat treatment of the cast ingot is, therefore, very important if the optimum dispersion of grain pinning particles is to be achieved. Figure 5 illustrates the very fine nature of the optimum distribution.

After homogenisation the ingot is hot rolled, usually on a reversing mill that also provides the opportunity for widening the ingot by cross rolling. In the majority of modern plants the hot flat blank that results from the reversing breakdown mill passes, without further heat treatment, into a multi stand hot strip mill after which it is coiled in readiness for subsequent cold rolling. The physical procedures used in hot and warm rolling the Supral alloys are no different from those employed for conventional strong aluminium alloys although precise control of temperatures and times is necessary in order not to allow excessive growth of the Al_3Zr particles.

As mentioned above, even employing specially developed casting hardware and procedures, some separation of coarse Al_3Zr occurs during solidification. If all of the rolling operations are uni-directional the coarse Al_3Zr particles become aligned leading to mechanical fibreing of the product and anisotropy of superplastic properties. This fibreing can be reduced by cross rolling although, when carried out during the cold rolling operation, this makes the rolling operation appreciably more expensive. The other respect in which the cold rolling of the Supral alloys differs from cold rolling of conventional alloys is in the extent of cold rolling without intermediate annealing. Conventionally, a normal strong alloy would be inter-annealed either after every pass or after every other pass (a reduction of, perhaps, 40-50%). The Supral alloys are designed to recrystallise dynamically as superplastic straining takes place and this dynamic recrystallisation is encouraged if the sheet, before superplastic deformation commences, is in a heavily cold worked condition. Intermediate annealing is, therefore, avoided and cold rolling reductions of the order of 70% are employed for optimum superplastic behaviour.

The Supral alloys, therefore, go into the superplastic forming operation in a heavily cold worked condition and still consisting, essentially, of the (relatively coarse) grain structure with which they were cast. When the correct structure has been obtained heating to the temperature of forming does not result in recrystallisation but subsequent plastic deformation in the forming operation produces a fine, recrystallised grain structure.

High Strength Aluminium Alloys

In developing a superplastically deformable version of established aircraft structural alloys, such as 7475, it was a basic objective at the outset to retain, unaltered, as much as possible of the manufacturing sequence in order to gain ready acceptance of the superplastically formed alloy in aerospace structures. In particular, the composition was intended to remain within the already established limits.

Thus, contrary to the Supral alloys, the early stages of melting, alloying, casting, homogenising and hot rolling would be identical with the same operations for non-superplastic 7475 strip. This has obvious advantages when it comes to casting although, it should be borne in mind that the high strength Al-Zn-Mg-Cu series is, certainly, amongst the most difficult series of aluminium alloys to cast.

The basic differences in manufacturing route compared with conventional, come into effect once hot rolling has been completed. Figure 6 indicates the route proposed by Rockwell([5 6]). In essence the strip, after hot rolling to approximately 25 mm is solution treated and then overaged to produce a dispersion of a very large number of particles with a size of about 0.5 μm. The metal is then heavily rolled at an intermediate temperature (approximately 85% at 200°C) before being given a recrystallisation treatment in a salt bath. The dispersion of particles in the heavily worked matrix results, on rapid heating, in the particles acting, in effect, as recrystallisation nuclei. Although by no means all of the particles act as nuclei, a reasonably fine grain size can be achieved by this processing route and good superplastic behaviour can be obtained. The superplastic behaviour achieved is, very much, related to the grain size achieved and Figure 7 ([10]) demonstrates both the greater superplastic ductilities and higher strain rates made possible by grain size refinement to about 5 μm. However, while the above processes are possible in a laboratory their direct translation into a factory environment would be very difficult. Considering the Rockwell sequence of Figure 6, reference to Figure 2 will show that the natural point at which to solution treat would be after hot strip rolling. At this stage the strip has been coiled and the rapid cooling implicit in solution treatment would not be possible in coil form. If it is assumed that this difficulty can be overcome, the next step in the Figure 6 sequence is a very large warm rolling reduction. This too, would be very difficult to achieve with conventional production equipment, particularly in thinner (< 3 mm) gauge material. The final stage of salt bath annealing is a perfectly practicable step but very few large aluminium companies are now equipped for this operation.

In practice, at least four major aluminium companies are manufacturing superplastically deformable versions of 7475 sheet so it must be assumed that there are ways around the obstacles mentioned above. Little, however, has been published on the topic by the Companies concerned although there is a series of patents which allude to the problems of conducting the sequence of Figure 6 in a production unit and which suggest modifications to allow a more industrially practicable route. Ward, Agrawal and Ashton ([11]) demonstrate that after solutionising the hot rolled blank it is not essential to water quench (or otherwise rapidly cool) before the overageing treatment and, indeed, they claim a superior final product by cooling directly from the solutionising temperature to the overageing temperature. Having achieved the appropriate size and distribution of precipitates the material can then be further rolled either at elevated temperature or at ambient temperature. Where thicker gauges are required the overageing is applied to a relatively thick blank which is then either warm or warm and cold rolled to gauges in the range 3 mm to 1.5 mm. As with Supral alloys it is suggested that it is advantageous to cold roll transversely to the hot/warm rolling direction. For thinner gauges the hot rolled blank is overaged when it is about 6.25 mm thick and then cold rolled, in coil form, to the intended gauge.

A further development of this thinking is described by Ward and Ashton([12]) and shown schematically in Figure 8(b). In this case both water quenching after solutionising and the discreet overageing treatment are eliminated. The part rolled, solutionised slab is cooled at a controlled rate (ideally ~ 25°C per hour) to the temperature of hot rolling and hot rolling can then commence without further delay. While earlier patents have restricted the temperature of rolling to about 275°C (i.e. warm not hot rolling), material being produced by this route can, initially, be hot rolled although the temperature will fall into the warm working range as rolling proceeds. During this warm working the equivalent of some cold work will be imparted to the strip but it is, apparently, important that the strip temperature should be below about 275°C at the end of the hot/warm rolling operation. The coil, typically at a gauge of about 7.5 mm, would then be cold rolled to final gauge.

A further development of the above ideas([13]), shown schematically in Figure 8(c), incorporates the alternatives of either cold or warm rolling to impart sufficient stored energy to produce the fine recrystallised structure. It is claimed that this technique produces a strip that can be recrystallised to the required structure using the heating rates available in continuous heat treatment furnaces.

A further method of improving the manufacturing route is described by Bampton ([14]) who points out that the overaged material produced by the initial Rockwell route continues to age naturally at ambient temperature once the overageing treatment is completed. This results in cracking during cold rolling. If the overaged material is cold rolled without delay - which in industrial production terms will frequently be impractical - the relatively high level of supersaturation will also result in cracking after relatively small rolling strains. Bampton shows that a further softening treatment, after completion of the overageing, by slowly cooling to a lower temperature (of the order of 250°C) and holding for about 24 hours removes both supersaturation and GP zone. The strip is then sufficiently ductile to be cold rolled to larger strains and finer grain sizes are claimed to be possible. While this technique was designed for application to material that had been separately overaged, it would appear to be equally applicable to material processed according to the ideas expressed in the patents of references ([11 12 13]).

It has also been pointed out ([15]) that the rate of cavitation in fine grained 7475 can be significantly reduced if the sheet is held at, or near, the superplastic forming temperature prior to forming. It is suggested that the pre-anneal reduces the likelihood of incipient melting of the two phase constituent particles and allows reduction of the size of H_2 pores. The treatment is said to be more effective with material that has a high cavitational tendency and can, perhaps, be regarded as an expedient step to correct earlier, non-optimum stages in the manufacture of the sheet.

Aluminium-Lithium Based Alloys

Probably the largest single development in the aluminium companies at present, is concerned with the production of aluminium-lithium based alloys. These alloys, containing between ~ 1.75% and ~ 2.75% of lithium are of great interest for aerospace applications because they hold out the possibility of reducing the density of the alloy by between 6 and 12% while, simultaneously achieving modulus increases of the same order. The combined pressures from the ever-present search for higher performance in military aircraft, together with the wish to make civil aircraft more fuel efficient and the threat to aluminium posed by the development of high performance, carbon fibre reinforced composites resulted in the current major developments in the aluminium companies, the aerospace constructors and universities and research institutions.

The major aluminium company developments all have very similar objectives. Essentially, the new alloys are used to replace the established alloys used in airframe construction and the alloy developers are seeking to match the service properties of the established aerospace alloys, while gaining density and stiffness improvements. The targets vary somewhat, particularly in terms of target density reduction, but basically all programmes seek to produce a damage tolerant alloy, suitable for 2024 T3 replacement, a medium strength, general purpose alloy to match 2014 T6 properties and a high strength alloy to match the properties of the high strength, 7000 series alloys. Work has been carried out on alloys from a number of sub-families:

Al-Mg-Li

Al-Cu-Li

Al-Li-Cu-Mg

all of which contain a zirconium addition of about 0.1%. However, the likely alloys for initial commercial use are now becoming clearer and it has been demonstrated that early fears that the new alloys would not match the baseline alloys in terms of toughness have been shown to be unfounded. At least three demonstrator aircraft make some structural use of the new aluminium-lithium based alloys and several demonstration components have been superplastically formed by British Aerospace from Alcan Lital 8090. Some of the contending new alloys are listed in Table 3.

However, most of the published work concerning superplastically deformable aluminium-lithium sheet has reported the results of laboratory experiments and only very limited activity has involved material produced on a commercial scale. At least two of the major aluminium companies involved in the aluminium-lithium development (Alcan and Alcoa) have stated their intention of developing superplastic quality sheet but none is yet available commercially. Thus, it is not possible in this section to describe the manufacturing routes for superplastic quality aluminium-lithium alloy sheet.

Nevertheless, it is already possible to make a series of observations that is likely to have repercusions in the eventual commercial manufacturing route. Firstly it is clear that many (see Table 4 (16)), if not all, of the aluminium-lithium based alloys now being proposed for aircraft structural applications can be deformed superplastically, if processed appropriately. Thus high zirconium additions, or their equivalent, as in the Supral alloys, are not necessary and the casting operation need be no different from that for other aluminium-lithium alloy products. It should, however, be noted that, because of the high reactivity of aluminium-lithium based alloys in the molten state special casting techniques have to be employed, regardless of the end product form. Great care over furnace atmosphere control is necessary if the rapid loss of lithium by oxidation is to be avoided. Use of the conventional reverberatory furnaces normally employed in the aluminium industry cannot be allowed and special furnace refractories must be employed as the molten alloys rapidly attack many conventional refractories. It has also been demonstrated that molten aluminium-lithium alloys can in certain circumstances, react explosively with water, so that special casting procedures are essential for safety reasons.

It can be seen from Table 4 that it is possible to induce superplastic behaviour in aluminium-lithium based alloys either by the use of a "Rockwell" processing route to yield a fully recrystallised fine grained, structure before commencing superplastic deformation(17) or by employing a "Supral" processing route(18) in which the starting sheet is in a wrought condition before superplastic deformation commences and a fine grain size is generated simultaneously with the forming operation. Both routes are capable of providing large superplastic strains and insufficient evidence is probably currently available to enable a confident judgement as to which route has the greatest superplastic potential. Additionally, several workers have shown superplastic behaviour in sheet made via a powder metallurgy route and quite good ductilities have been reported(19). However, from the viewpoint of a metal manufacturer the Supral route is appreciably more attractive than either thermo-mechanical processing to a fine recrystallised grain structure or the powder route since neither the very large cold reductions (without annealing) are required and nor are the very high heating rates to give static recrystallisation. Certainly, if the Supral route is employed to manufacture the sheet and superplastic deformation takes place using an initial high strain rate to generate a fine recrystallised grain structure followed by continuing deformation at a lower strain rate(20) large superplastic strains can be achieved. If this dual strain rate technique is employed and the deformation takes place under superimposed hydrostatic stress, as recently reported by Ghosh et al (21), then superplastic strains can be achieved well in excess of any that are likely to be required in component manufacture in the immediately foreseeable future. This being so it seems unlikely that the use of the more complex "Rockwell" route could be justified and it is the writers belief that superplastically deformable aluminium-lithium, sheet made by the "Supral" route will be commercially available in the very near future.

This sheet is likely to be available in either the 8090 or the 8091 composition(22). The former is somewhat less superplastic but has lower quench sensitivity and it is therefore possible to cool relatively slowly after the superplastic forming operation so avoiding quenching distortion. The latter is more quench sensitive and a water quench may be necessary in order to achieve solution treatment. However, the 8091 is capable of achieving appreciably higher strengths than the 8090. The 8091 alloy is more difficult, from an aluminium manufacturers' viewpoint because both casting and cold rolling give more problems than the lower strength alloy 8090.

Like titanium alloys, some superplastic behaviour can be obtained from much "standard" aluminium-lithium sheet. To obtain this behaviour the sheet needs to be in a non-recrystallised condition and difficulties then derive from the relatively high level of anisotropy in conventionally processed non-recrystallised sheet. Additionally, sheet not processed for the superplastic end use may often partially recrystallise either during solution treatment, at the sheet manufacturer, or during pre-heating for superplastic forming. In either event, recrystallisation will lead to severely reduced superplastic strain capability and serious cavitation in the vicinity of the coarse grains.

Eutectic Aluminium Alloys

Although no eutectic superplastic aluminium alloy has yet achieved significant commercial use several potentially useful alloys have been suggested and the possibility remains that an aluminium eutectic alloy will achieve more extensive use. Given the importance of eutectic alloys in the earlier development of superplastic alloys, it seems appropriate to include a brief mention of eutectic alloy manufacture in this review.

From the point of view of the metal manufacturer, the eutectic alloys are probably the most attractive. Special casting techniques are usually not required, although it would generally be wished to achieve as fine a mixture of the eutectic components in the cast product as possible. This would most readily be achieved by rapid solidification and would, therefore, be aided by a small ingot size. Probably the most developed of the aluminium based eutectics was the Al-Ca-Zn alloy described by Moore and Morris (23). In casting they required a growth coupled eutectic to form in which the Ca-Zn-Al intermetallic forms as small diameter rods. Thereafter, conventional hot and cold rolling results in break up of the eutectic structure so that the eventual structure is of Ca-Zn-Al particles of about 2 μm diameter in an aluminium matrix, but prolonged exposure to elevated temperatures must be avoided to prevent coarsening of the Ca-Zn-Al particles. The selected ternary composition allows satisfactory cold rolling whereas the binary Al-Ca eutectic (24) tends to be considerably more brittle and to be difficult to cold roll.

TITANIUM ALLOYS

In most of the aluminium alloys considered above, superplastic behaviour was achieved either by a dynamic recrystallisation route or by thermal mechanical processing to provide a fine recrystallised starting grain structure but, in either case, producing materials that are, essentially, single phase (fcc) at the temperature of superplastic forming. Eutectic aluminium alloys contain, depending upon the particular eutectic, two phases present in approximately equal volume. The relative volumes are, however, largely determined at the casting stage. With titanium a quite different situation exists since the metal undergoes an allotropic modification at 882°C. Below this temperature the close packed hexagonal alpha phase dominates while above this temperature the body centred cubic beta modification forms.

Superplastic behaviour in titanium alloys was first reported (25) 20 years ago and it has since become clear that many titanium alloys can exhibit, at least some, superplastic behaviour. Table 5, derived from Hamilton (26) illustrates this point. Although both Hamilton (26) and Hammond(27) make the points that the mechanisms of superplastic deformation are incompletely understood and that scope exists for further development of superplastically deformable titanium alloys, the fact remains that titanium alloys seem to have a relatively high inherent superplastic forming capability after conventional manufacture. In particular, as can be seen from Table 5, the Ti-6Al-4V is capable of good superplastic deformation, has a fairly high strain rate sensitivity (m value) and exhibits superplastic behaviour at rather higher strain rates than the other alloys listed. The Ti-6Al-4V alloy also possesses very attractive service properties from the point of view of the airframe constructor, (minimum 0.2% PS 830 MPa, TS typically 960 MPa together with good corrosion, welding and diffusion bonding behaviour while no post forming heat treatments are required) and so has gained widespread acceptance as an aerospace structural material. Given this acceptance for structural use and the very reasonable superplastic behaviour, it can be seen that there is really not a very large incentive to develop new superplastic titanium alloys. Because the pure metal is very soft aluminium and vanadium are added to make the Ti-6Al-4V alloy. The aluminium acts as a stabiliser of the alpha phase while vanadium stabilises the beta phase. At ambient temperature the alloy is virtually all alpha phase although the form of the alpha will be dependent upon the rate of cooling, water quenching giving acicular alpha and fast air cooling fine needles. At about 900°C-925°C the structure is approximately 50% alpha : 50% beta. Numerous suggestions have been made to improve the superplastic behaviour of titanium, particularly including the additions of beta stabilising elements such as iron, nickel or cobalt to *depress the forming temperature(28). The addition of rare earth oxides, such as yttria, has been* suggested in order to stabilise the fine grain structure during superplastic forming and has been shown to be effective (29). However, all such changes have the disadvantage of leading to alloys outside the accepted chemistry of the established alloys with the consequent need to qualify all aspects of the modified alloy for aerospace use.

The basic manufacturing route for titanium and titanium alloys differs fundamentally from that for aluminium, and the great majority of commonly used metals, in that at no point does a furnace full of molten alloy exist. The starting metallic material is titanium sponge derived by reduction of the titanium ore (largely rutile or ilmenite containing, respectively, 50-60% TiO_2 and 95% TiO_2). However, because of the very high reactivity of molten titanium and its attack of refractory materials all melting has to take place either in vacuum or an inert atmosphere. Rather than melt a sufficient quantity of sponge from which to cast an ingot the sponge is melted in a consumable electrode arc melting furnace in vacuum. With this system the electrode is made from consolidated sponge after blending with the appropriate alloying additions. The electrode is melted in a cooled chamber (cooling being either by means of water or liquid NaK; the advantage of the latter is that it does not react with molten titanium, so that the explosion hazard that can exist with water cooling, if the molten alloy leaks through the crucible, is eliminated) and, at any given moment, only a limited amount of liquid metal (or alloy) exists. It is, therefore, very important, when making titanium base alloys, to achieve very accurate and uniform mixing of the alloy components in the electrode since conventional mixing of the bath cannot take place and nor, of course, can alloy corrections be made to the bath. For quality improvement, it is frequently the practice to re-melt the ingot a second time. While this may give the impression of a relatively small scale operation, the technique is applied to produce ingots of 1250 mm, or so, diameter weighing of the order of 17 tonnes [30].

Conversion of the titanium alloy ingot into sheet takes place by a reasonably conventional route. The alloys can be hot rolled readily and, provided that the ingot pre-heating occurs in an atmosphere controlled furnace (oxidising, because hydrogen pick-up can occur in a reducing atmosphere and this can lead to embrittlement) reasonably conventional furnaces can be used. The basic objective is to produce a sheet material with a very fine grain size and, preferably, reasonable freedom from crystallograhic preferred orientation. While Hamilton[26] clearly demonstrates a decrease in value from > 0.9 in Ti-6Al-4V with a grain size of 6.4 μm to < 0.7 when the grain size increases to 20μm. Peters et al[31] found that special processing to give an alpha grain size of ~ 1.2 μm actually resulted in a lower m value than the conventionally processed sheet. It was concluded that disadvantages from textural effects were overriding the benefit from the very fine grain size as far as m value (and maximum strain capability) were concerned. Nevertheless, the fine grained material gave lower flow stresses than the "conventional" material which, from a practical forming viewpoint would frequently be more important than the maximum strain capability. Partridge et al [32] investigated causes of superplastic anisotropy in Ti-6Al-4V and concluded that the anisotropy of superplastic deformation derived not from crystallographic preferred orientation but from the presence of contiguous alpha phase. The aligned, stronger, alpha phase presented barriers to grain boundary sliding, hence resulting in anistropic flow.

Overall, therefore, the conventional production route for the conventional Ti-6Al-4V alloy, with hot working in the alpha plus beta region followed by finish rolling at temperatures down to about 800°C resulting in a refined alpha phase and high uniformity of bi-modal structure gives a material capable of very good superplastic behaviour.

CONCLUSIONS

This review has only considered the manufacture of a very limited number of superplastically deformable materials when very large numbers of such materials have been shown to exist. Nevertheless, it has attempted to demonstrate the important metallurgical processing variables that have to be considered when manufacturing any superplastic alloy despite the very limited information in the public domain. In the case of the "Supral" aluminium alloys, the degree of success achieved in producing a good superplastic product is, very largely, determined at the casting stage. For virtually all the other alloys it is the details of thermal/mechanical processing that determine the superplastic quality of the product. Given the success that can be achieved in inducing superplastic behaviour in a very wide range of "standard" alloys there can now be very little justification in modifying alloy compositions simply to improve superplastic behaviour.

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TABLE 1

EARLIER* ALLOY SYSTEMS EXHIBITING SUPERPLASTIC BEHAVIOUR

Alloy System	Superplastic temperature (°C)	m	Elongation (%)
Al-Cu (eutectic)	480	0.9	500
Al-Cu-Zr (Supral)	460	0.6	1200
Cd-Zn	20	0.5	400
Co-Al	1200	0.5	850
Cu-Zn	500	0.5	300
Cu-Al-Fe	800	0.5	720
Fe-Cr-Ni	920	0.5	600
Mg-Zn-Zr	300	0.6	1000
Mg-Al (eutectic)	375	0.8	2100
Ni-Cr-Fe-Ti	980	0.5	1000
Sn-Bi	20	0.5	1000
Ti-Al-V	920	0.85	450
Zn (commercial)	60	0.2	400
Zn-Al (eutectic)	270	0.7	1500

TABLE 2

NEWER ALLOY SYSTEMS EXHIBITING SUPERPLASTIC BEHAVIOUR

Alloy System	m value	Elongation (%)
Al-Cu-Mg-Ge-Zr (Supral 220)	0.5	1200
Al-Zn-Mg-Zr	0.5	1200
Al-Zn-Mg-Cu (AA 7475)	0.85	1000
Al-Ca	0.78	850
Al-Ca-Zn	0.5	600
Al-Li-Cu-Mg (8090)	0.6	1000
Al-Cu-Li (2090)		
Fe-Cr-Ni (IN 744)	0.5	1000

TABLE 3

ALUMINIUM-LITHIUM BASED ALLOYS FOR AEROSPACE

Alloy	Composition (wt %)			
	Li	Cu	Mg	Zr
Alcan Lital 8090	2.4	1.2	0.6	0.1
Alcan Lital 8091	2.5	1.8	0.7	0.1
Pechiney (CP274) 2091	2.0	2.1	1.5	0.1
Pechiney CP 276	2.3	2.9	0.4	0.1
Alcoa Alithalite 2090	2.2	2.7	-	0.1

TABLE 4

ALUMINIUM LITHIUM BASED ALLOYS EXHIBITING SUPERPLASTICITY

Composition (wt %)				Elongation (%)
Li	Cu	Mg	Zr	
2.0	-	-	-	320
3.3	-	-	0.15	340
3.0	-	-	0.5	1035
2.7	-	2.8	0.15	680
2.5	1.2	0.5	0.10	875
3.0	4.0	-	0.5	825
1.9	2.9	1.0	0.15	798
1.9	2.8	0.9	0.2	654
2.0	4.0	-	0.2	700
2.5	1.8	0.7	0.12	1200

TABLE 5

SUPERPLASTIC TITANIUM ALLOYS

Alloy	Temperature (°C)	Strain Rate (sec $^{-1}$)	m	Elongation (%)
Ti-6AL-4V	920	5×10^{-4}	0.75	1100
Ti-6Al-5V	850	8×10^{-4}	0.70	1100
Ti-6Al-2Sn-4Zr-2Mo	900	2×10^{-4}	0.67	540
Ti-6Al-4V-2Ni	815	2×10^{-4}	0.85	720
Ti-6Al-4V-2Co	815	2×10^{-4}	0.54	650
Ti-5Al-2.5Sn	1000	2×10^{-4}	0.49	420

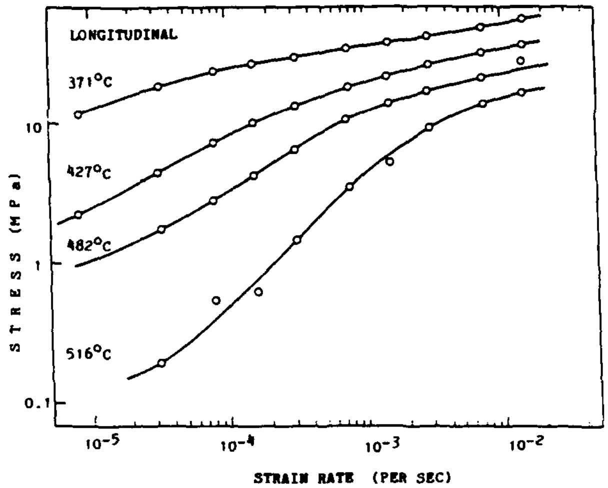


FIGURE 1: VARIATION IN FLOW STRESS WITH TEMPERATURE AND STRAIN RATE FOR 7475 (AFTER HAMILTON ET AL[5])

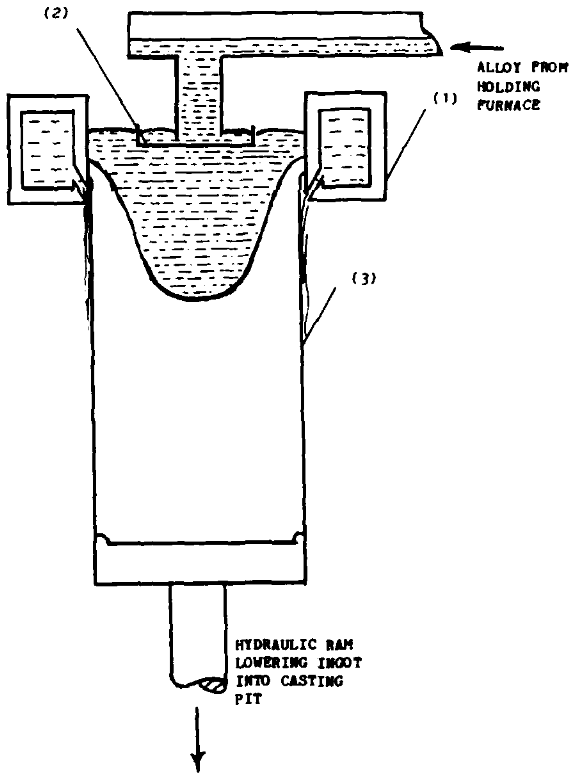


FIGURE 2: SCHEMATIC OF ALUMINIUM SHEET PRODUCTION ROUTE (A) THE VERTICAL SEMI-CONTINUOUS DIRECT CHILL CASTING SYSTEM

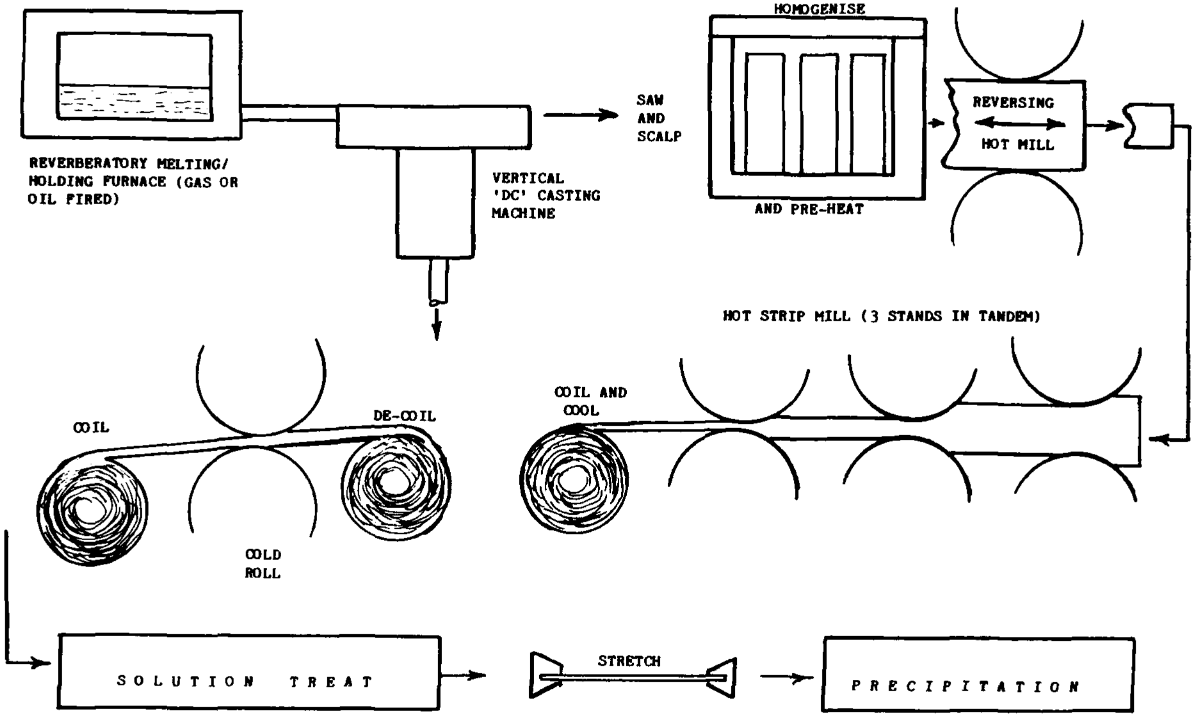


FIGURE 2: SCHEMATIC OF ALUMINIUM SHEET PRODUCTION ROUTE
(B) THE FABRICATING SEQUENCE

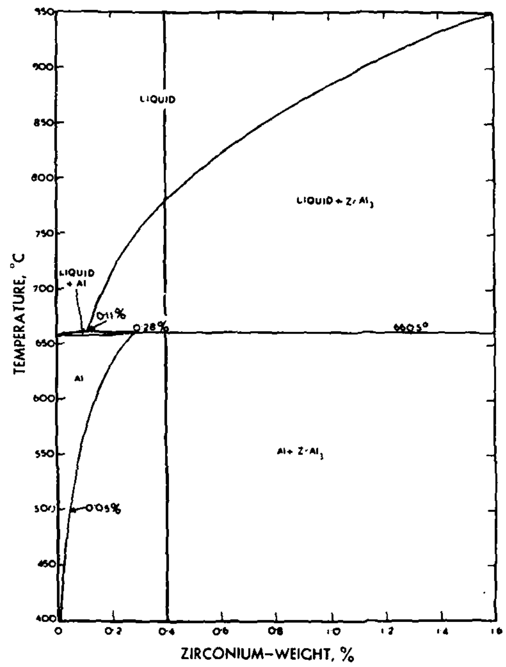


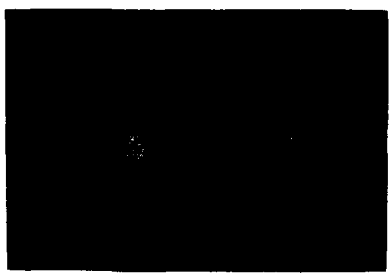
FIGURE 3: THE ALUMINIUM-ZIRCONIUM EQUILIBRIUM DIAGRAM (AFTER H.W.L. PHILLIPS. INST. METALS MONOGRAPH SERIES No.25)

(A)

(B)

X1000

FIGURE 4: THE DIFFERENT FORMS IN WHICH COARSE Al_3Zr CAN APPEAR IN THE MICROSTRUCTURE OF CAST SUPRAL ALLOYS



X13000

FIGURE 5: DISTRIBUTION OF FINE SCALE Al_3Zr FOR PINNING GRAIN BOUNDARIES

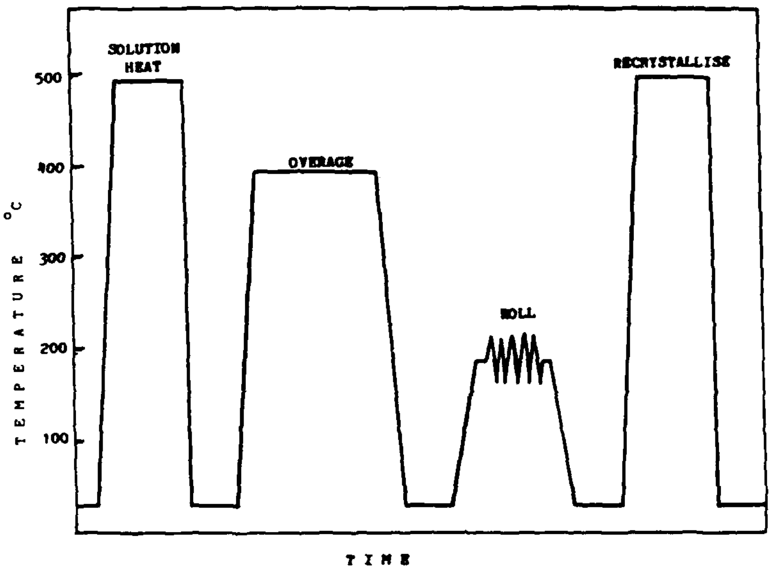


FIGURE 6: SCHEMATIC OUTLINE OF ROUTE FOR THE PRODUCTION OF FINE GRAINED 7475 (AFTER HAMILTON ET AL[5])

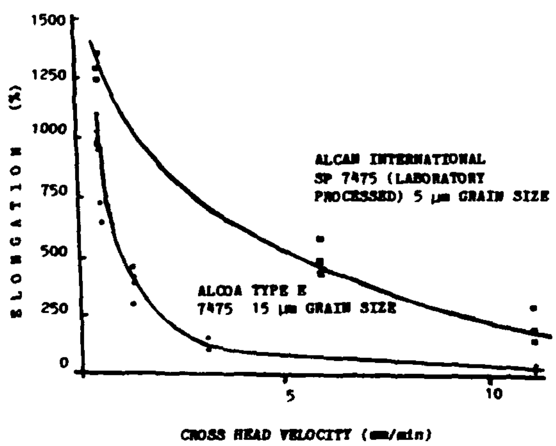


FIGURE 7: INFLUENCE OF GRAIN SIZE ON SUPERPLASTIC PERFORMANCE OF 7475 SHEET

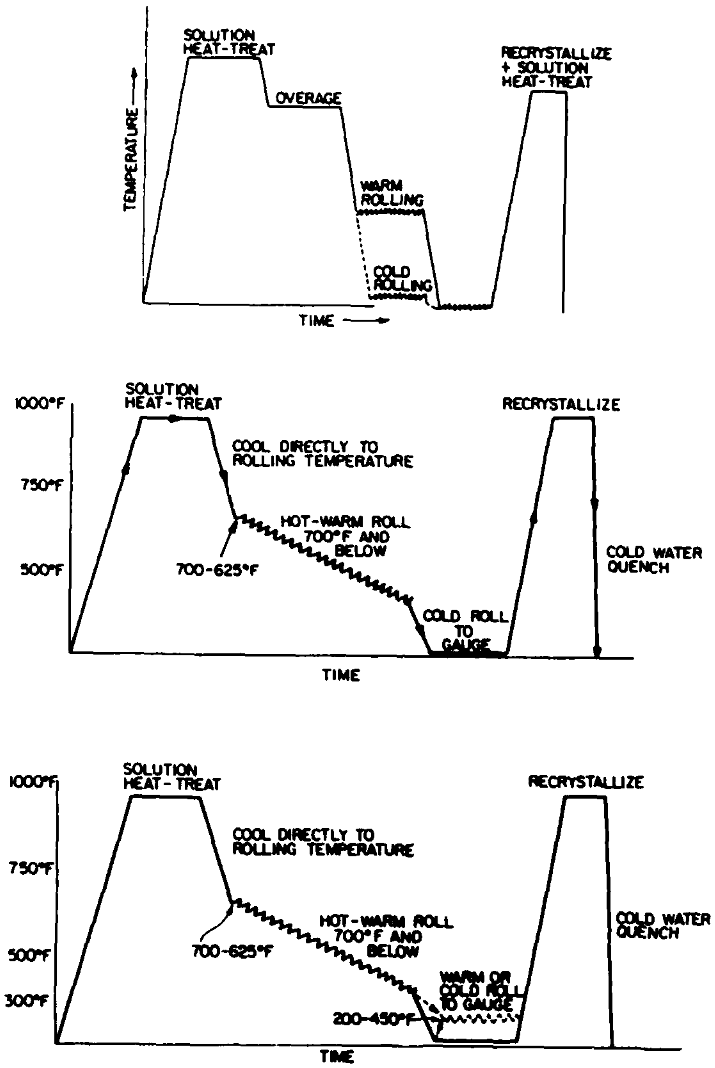


FIGURE 8: SCHEMATIC PROCESSING ROUTES FROM US PATENTS ILLUSTRATING THE DEVELOPMENT OF PRACTICABLE COMMERCIAL SCALE ROUTES FOR SUPERPLASTIC QUALITY 7475 SHEET.

BIBLIOGRAPHY

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AD-A203 614/3/XAD NTIS
Effect of Superplastic Deformation on the Surface Roughness of Sheet
Partridge, P. G.; Dunford, D. V. Royal Aerospace Establishment, Farnborough (England).*Defence Research Information Centre, Glasgow (UK) (092599000 419293) Technical memo, Rept no: RAE-TM-MAT/STR-1112, Jul 88, 8p, Monitor: DRIC-BR-108126, NTIS Prices: PC A02/MF A01

SPF/DB 1 involves in situ diffusion bonding after the sheet has undergone large superplastic strain. Bond quality depends primarily on precise process control which is based upon the forming and bonding parameters. Surface roughness is one of the most important parameters affecting solid state bonds since in practice increased surface roughness requires an increase in the bonding time. Plastic deformation is known to increase the surface roughness of CP titanium sheet, but although an increase in surface roughness has been reported during superplastic deformation of Titanium-alloy sheet, no roughness measurements were made. Ra values have therefore been determined for titanium and aluminum alloy sheets before and after SPF and are reported in this paper. Their significance for forming and bonding of sheet is discussed. (jes)

Fld: 71N, 41F *Controlled terms:* *Aluminum alloys/*Diffusion bonding/*Plastic deformation/*Titanium/Bonding/Control/Measurement/Parameters / Precision/Quality/Roughness/Sheets/Surface roughness/Time / Uncontrolled terms: *Foreign technology/*Superplastic metal deformation / NTISDODXA/NTISFNUK

89A28432 NASA IAA Journal Article Issue 10
Structural materials for future aerospace developments. Materiales estructurales para los futuros desarrollos aeroespaciales
(AA)GARCIA POGGIO, JOSE A. (Congreso Nacional de Ingenieria Mecanica, 6th, Madrid, Spain, Dec. 15-18, 1987) Ingenieria Aeronautica y Astronautica (*ISSN 0020-1006*), *no. 309, 1989, p. 8-20. In Spanish.* 890000 p. 13 refs 54 In: SP (Spanish) p.1453

A comprehensive evaluation is presented of the state-of-the-art and prospective developments in high strength/weight and heat-resistant materials applicable to the primary and secondary airframe structures of such advanced aerospacecraft as those of tilt-rotor VTOL, high-altitude solar propulsion, SST, and hypersonic cruise type; large, complex satellites, OTVs, and the manned NASA Space Station are additional fields of application. The materials discussed are Al-Li alloys, ceramic fiber-reinforced metal-matrix composites, high service temperature Al-Cr-Fe alloys produced by vapor-phase condensation, mechanically alloyed Al alloys, ARALL laminates, and superalloys and ceramics for propulsion-system applications. O.C.

Category code: 23 (chemistry/materials) Controlled terms: *AIRCRAFT CONSTRUCTION MATERIALS /*HEAT RESISTANT ALLOYS /* HIGH STRENGTH ALLOYS /*SPACECRAFT CONSTRUCTION MATERIALS / ALUMINUM ALLOYS/LAMINATES/LITHIUM ALLOYS / SUPERPLASTICITY/USER REQUIREMENTS /

AD-A202 543/5/XAD NTIS
Deformation of Electron Beam Welded Ti-6Al-4V Alloy Sheet under Superplastic Conditions
Partridge, P. G.; Dunford, D. V. Royal Aerospace Establishment, Farnborough (England).*Defence Research Information Centre, Glasgow (UK) (092599000 419293) Technical memo, Rept no: RAE-TM-MAT/STR-1109, Jul 88, 8p, Monitor: DRIC-BR-108269, NTIS Prices: PC A02/MF A01

Electron beam welds in Ti-6Al-4V sheet deformed under superplastic conditions were less superplastic and deformed less than the adjacent sheet. This caused severe local thinning of the sheet when the weld direction deviated from the principal strain direction. However, deformation removed weld undercuts and caused acicular weld microstructures to become more like the equiaxed sheet microstructure. The implications for the superplastic forming of components from welded sheet, bar and extrusions are discussed. Great Britain. (JES)

Fld: 71N Controlled terms: *Deformation/*Welds/*Sheets/Electron beams/ Extrusion/Microstructure/Deformation/Electron beams/Extrusion / Microstructure/Uncontrolled terms: *Foreign technology/Titanium/Aluminum / Vanadium/*Plastic properties/NTISDODXA/NTISFNUK

55-1356 METADEX Issue 8906
Diffusion Bonding Titanium Structures for Aeroengine Components.
Fitzpatrick, G. A. Rolls-Royce Diffusion Bonding; Cranfield, UK; 7-8 July 1987; Cranfield Institute of Technology, School of Industrial Science; Cranfield, Bedford MK43 0AL, UK; 1987; 8906-72-0288; 253-258; in English

Solid-state diffusion bonding, a joining process which is being actively developed by Rolls-Royce for the manufacture of Ti fabrications, sometimes in conjunction with superplastic forming, is described. Examples to date include panel structure with varying internal configurations, and hollow vane/blade constructions. Chemically machined surfaces, high applied pressures and elevated temperatures have been found necessary to effect a solid-state diffusion bond in these Ti alloy assemblies. As the understanding of the manufacturing control of solid-state diffusion bonding for Ti alloy components and its effect on bond performance is established, together with the development of suitable verification techniques whcch guarantee product integrity, then Rolls-Royce will utilise the process in the manufacture of several aeroengine components. Applications already identified include lightweight fabrications, particularly where SPF/DB techniques are employed in conjunction with other advanced fabrication processes and/or advanced materials.—AA

Category: 55 Controlled Terms: Titanium base alloys/Bonding/Diffusion welding/Aerospace engines/Superplastic forming/Liquid phases

AD-A201 732/5/XAD NTIS
Deformation of Extruded Titanium Alloys Under Superplastic Conditions
Dunford, D. V.; Partridge, P. G. Royal Aerospace Establishment, Farnborough (England).*Defence Research Information Centre, Glasgow (UK) (092599000 419293) Technical memo, Rept no: RAE-TM-MAT/STR-1111, Jul 88, 9p, Monitor: DRIC-BR-108112, NTIS Prices: PC A02/MF A01

The potential cost and weight savings associated with superplastic forming of thin Ti-6A1-4V sheet structures (1) has led to increased interest in the forming of other product forms such as bar and extruded sections. However, compared with thin sheet, extruded Ti-6A1-4V alloy has a coarser and less uniform microstructure which would tend to be less superplastic than sheet. The M-values, microsrructure and deformation behaviour of an extruded U-channel section in IMI 550 and IMI 318 has been studied under conditions that produce superplasticity in these alloys in thin sheet form. The results are compared with those obtained for bar and sheet. Great Britain. (jes)

Fld: 71N Controlled terms: *Alloys/*Extrusion/*Microstructure/*Plastic properties/Costs/Deformation/Sheets/Thinness/Titanium alloys/Weight reduction/Uncontrolled terms: *Foreign techoology/NTISDODXA/NTISFNUK

46-0118 METADEX Issue 8906
Titanium: the Right Stuff for Land, Air and Sea.
Hunt, M. Mater. Eng. (Cleveland) Oct. 1988; 105, (10); 45-49; ISSN 0025-5319; in English

The recent developments of Ti (e.g. Ti-6Al-4V, Ti-6242S, Ti-1100, Ti-3Al-2.5V, Ti-3Al-8V-6Cr-4Zr-4Mo) metallurgy are discussed in terms of ingot casting, forging, rapid solidification and powder metallurgy for applications in the automotive, aerospace, marine and energy industries. Further improvements are outlined, which include hot isostatic pressing of investment casting for engine frame, thermochemiaal heat treatment to refine microstructures and development of tool materials to overcome the machining difficulty.-B.C.

Category: 46 Controlled Terms: Titanium base alloys/Alloy development / Sponge metal/Powder metallurgy/ Investment casting/Superplastic forming / Aircraft components/Materials selection Alloys: Ti-6Al-4V / Ti-6Al-2Sn-4Zr-2Mo-0.1Si/Ti6242S/Ti-3Al-2.5V/Ti1100/TI / Ti-3Al-8V-6Cr-4Zr-4Mo/TI

46-0109 METADEX Issue 8905
New Light Aluminium Alloys.
Ferton, D. Cegedur Pechiney 8th International Light Metals Congress (8 Internationale Leichmetalltagung); Leoben/Vienna, Austria; 22-26 June 1987; Auuminium-Verlag GmbH; Konigsallee 30, D-4000 Dusseldorf 1, FRG; 1987; 8905-72-0247; 738-743; in English

Conventional Al alloys can be still potentially improved, especially their engineering properties. This is particularly true of their superplastic forming capacity. Al-Li alloys provide a 15% weight saving while meeting the other properties required for aerospace applications. Aluminium powder-made alloys become extremely interesting at room and elevated temperatures. Aluminium metal matrix composites represent an important departure from the development of Al alloys, with a 100% increase in some properties. There is no doubt that Al and new Al alloys, possibly combined with other materials, will remain the major component of numerous applications. 13 ref.-AA

Category: 46 Controlled Terms: Aluminum base alloys/Alloy development/Powder metallurgy/Dispersion hardening alloys/Composite materials/Alloying elements Alloys: 7075/Al-2.5Li-11Be 2091/2214/CP276/AL

89N16034 NASA STAR Technical Report Issue 08
Superplastic forming of 8091 aluminum lithium
Report, Apr. 1987 — Jun.1988 (AA)ANTONE, C. E.; (AB)MARTIN, G. R.; (AC)GHOSH, A. K.; (AD)GANDHI, G Rockwell International Corp., Los Angeles, CA. (RY232014) Aircraft Div. AD-A200364; AFWAL-TR-88-3074 F33615-87-C-3223 880621 p. 18 In: EN (English) Avail: NTIS HC A03/MF A01 p.1051

Aluminum-Lithium alloys have been introduced to the Aerospace community as a way to decrease weight and improve stiffness over conventional aluiinum alloys for structural components. A manufacturing method which as created a great deal of interest for Al-Li aerospace applications is the fabrication of net shape parts by superplastic forming (SPF). Aluminum-Lithium alloys present some unique handling problems and fabrication challenges for established practices in superplastic forming. This paper will discuss the manufacturing challenges and approaches of forming 8091 AlLi by SPF and provide a brief overview into the material characteristics which make 8091 a successful candidate for SPF aircraft parts. GRA

Category code: 26 (metallic materials) Controlled terms: *ALUMINUM ALLOYS /*FORMING TECHNIQUES / *LITHIUM ALLOYS /* MANUFACTURING /*SUPERPLASTICITY / AEROSPACE SYSTEMS/AIRCRAFT PARTS/FIGHTER AIRCRAFT/STIFFNESS/WEIGHT REDUCTION/

89A15051 NASA IAA Meeting Paper Issue 03
Superplasticity in aerospace
Proceedings of the Topical Symposium, Phoenix, AZ, Jan. 25-28, 1988 (AA)HEIKKENEN, H. CHARLES; (AB)MCNELLEY, TERRY R. (AA)ED.; (AB)ED. (AA)(General Dynamics Corp., Fort Worth, TX); (AB)(U.S. Naval Postgraduate School, Monterey, CA) Symposium sponsored by the Metallurgical Society of AIME. Warrendale, PA, Metallurgical Society, Inc., 1988, 381 p. For individual items see A89-15052 to A89-15071. 880000 p. 381 In: EN (English) Members, $49.; nonmembers, $90 p.292

The present conference on emerging applications of superplastically formable materials discusses the superplastic deformation mechanisms of aluminum alloys, microstructural evolution in a P/M 7xxx Al alloy, fine-grained superplasticity at 300 C in a wrought Al-Mg alloy, and the extended ductility of Al and alpha-Fe alloys through dynamic recovery mechanisms. Also discussed are models for superplastic forming under axisymmetric and plane-strain conditions, cavitation failure in superplastic alloys, the effects of void formation on superplastic 7475 Al alloy mechanical properties, superplasticity in SiC whisker-reinforced Al alloys and mechanically-alloyed Al aerospace alloys, the effect of microstructure on Al-Li-X alloys' superplastic behavior, and hollow Ti-alloy turbofan blades' superplastic forming. O.C.

Category code: 26 (metallic materials) Controlled terms: *AIRCRAFT CONSTRUCTION MATERIALS / *CONFERENCES /*FORMING TECHNIQUES /*SPACECRAFT CONSTRUCTION MATERIALS /*SUPERPLASTICITY/ALUMINUM ALLOYS/DIFFUSION WELDING/FAILURE MODES / HEXAGONAL CELLS/LITHIUM ALLOYS/METAL WORKING/MICROSTRUCTURE/PLASTIC DEFORMATION/TURBINE BLADES /

AD-A200 732/6/XAD NTIS
International Conference on Superplasticity and Superplastic Forming
Hamilton, C. H.; Paton, N. E. Washington State Univ., Pullman.*Air Force Office of Scientific Research, Bolling AFB, DC (015416000 369850) Final rept. 1-4 Aug 88, 9 Aug 88, 23p, Grant: AFOSR-88-0158, Project: 2306 , Task: A1, Monitor: AFOSR-TR-88-1079. Prepared in cooperation with Rockwell International, Canoga Park, CA. NTIS Prices: PC A03/MF A01

It was apparent from the papers presented that the research and development activity in the area of superplasticity and superplastic forming is of substantial interest world-wide, and a number of papers reported results that are considered to be significant and which may point the direction for future research that should prove fruitful. Noteworthy among these are (1) the activities addressing high rate superplasticity, through both alloy development and process concept studies, (2) computer modeling of the SPF process, including finite element methods coupled with 3-D color graphics displays of thinning characteristics, (3) superplasticity in ceramic and intermetallic compound materials, (4) solid-state joining (diffusion bonding) of aluminum alloys, (5) demonstration that there are microstructural concepts other than that of fully recrystallized structure which can lead to superplasticity, especially at high rates, and (6) significant extension in the state of understanding of the interrelationship between microstructural dynamics and superplastic properties.

Aerospace equipment, Material forming, Superplastic forming, Airframes, Fabrication, Titanium alloys. (JES) Fld: 51C, 71D, 71N Controlled terms: *Aerospace systems/*Airframes/*Aluminum alloys/*Ceramic materials / *Computerized simulation/*Intermetallic compounds/*Materials / *Microstructure/*Plastic properties/*Titanium alloys/Addressing/Alloys / Diffusion bonding/Dynamics/Finite element analysis/High rate / International/Joining/Material forming/Solid state electronics / Symposia Uncontrolled terms: NTISDODXA/NTISDODAF

89A13001 NASA IAA Journal Article Issue 03
Principles of superplastic diffusion bonding
(AA)MAEHARA, Y.; (AB)KOMIZO, Y.; (AC)LANGDON, T. G. (AB)(Sumitomo Metal Industries, Amagasaki, Japan); (AC)(Southern California, University, Los Angeles, CA) NSF DMR-85-03224 Materials Science and Technology (ISSN 0267-0836), vol. 4, Aug. 1988, p. 669-674. 880800 p. 6 refs 50 In: EN (English) p.334

Superplastic diffusion bonding, which is growing in commercial/industrial importance in aerospace manufacturing, may occur in connection with either transformation or isothermal superplasticity. The nature of these two processes are presently characterized, with a view to their most advantageous applications. An account is given of the structural design features germane to superplastic forming in conjunction with diffusion bonding, for the case of such materials as Ti-6Al-4V. Since the bonding zone microstructure may undergo refinement as a result of dynamic recrystallization, overall properties are not degraded by the welding process. O.C.

Category code: 37 (mechanical engineering) Controlled terms: *DIFFUSION WELDING /*ISOTHERMAL PROCESSES /*PRESSURE WELDING /*SINTERING /*SOLID STATE /*SUPERPLASTICITY /FUSION WELDING / MECHANICAL PROPERTIES/MICROSTRUCTURE / PLASTIC DEFORMATION /

46-0049 METADEX Issue 8903
New Frontiers in Superlight Alloys.
Hunt, M. Mater. Eng. (Cleveland) Aug. 1988; 29-32; ISSN 0025-5319; in English

Al—Li alloys and magnesium alloys are being used in lightweight, high stiffness artificial limb components. These alloys were developed for the aerospace applications. Alcoa's 2090 is an Al—Li alloy with 7% lower density and 10% higher Young's modulus than 7075. This alloy is substituted for 7075 extrusions for McDonnell Douglas MD-11 wide cabin tri jet floor beams. Alcoa's X8092 can be substituted for 7075 plate. The X8092 is 8% lower in density and 9% higher in Young's modulus than the 7075. Kaiser's Al—Li alloys contain Mg and Cu for a better combination of strength, density, toughness, and corrosion. The alloy is available in extrusions, forgings, sheet, and plate. Allied-Signal is making up to 5% Li Al alloys with rapidly solidified powders. Zinc is added to reduce grain size. Table show the range of alloy compositions and the resulting strength, toughness and density. With the addition of Li in the alloys, the welding must be done to remove all hydrogen from the welding environment. Al—4Mg alloys are being developed for superplastic forming applications. The alloy ductility is limited by the formation of Al sub 8Mg sub 5 at the grain boundaries. Amax has improved the corrosion resistance of Mg by limiting the Cu and Ni to 0.25 and 0.01%. The 3.2% maximum for Mn provides the corrosion resistance. Dow and Allied-Signal are developing Mg—Si alloys for thixotropic injection molding (of granules) and rapidly solidified powders. A table details the properties of Mg casting alloys.—R.L.A.

Category: 46 Controlled Terms: Aluminum base alloys/Alloy development / Lithium/Alloying elements/Weldability/ Magnesium/Alloying elements / Magnesium base alloys/Alloy development/Density/Alloying effects/Tensile properties/Alloying effects/Toughness/Alloying effects/Superplasticity Alloys: 2090/7075/X8092/Kalite/905X6/Al-4Mg-1.3Li-1.1Cu-0.5O/9052 / Al-4Mg-1.1Cu-0.5O/Al-3.4Li-0.8Cu-0.4Mg-0.5Zr/AL/Mg-8.9Al-1.7Si-1.1Mn-0.8Zn / ZE41A/ZE33A/AZ91/AM60/AS41/MG

62-0074 METADEX Issue 8902
Superplasticity in Silicon Carbide Reinforced Aluminum and Mechanically Alloyed Aluminum Aerospace Alloys.
Chokshi, A. H.; Bieler, T. R.; Nieh, T. G.; Wadsworth, J.; Mukherjee, A. K. University of California, Lockheed Missiles and Space. Superplasticity in Aerospace; Phoenix, Arizona, USA; 25-28 Jan. 1988; The Metallurgical Society/AIME; 420 Commonwealth Dr., Warrendale, Pennsylvania 15086, USA; 1988; 8902-72-0074; 229-245; in English

The addition of silicon carbide, in either whisker or particulate form, improves the ambient temperature strength and stiffness of Al-based alloys. Recent studies have shown that, after appropriate thermo-mechanical treatment, it is possible to induce superplastic-like behavior in such composites. In particular, a silicon carbide whisker reinforced 2124 Al composite has been processed to exhibit large elongations to failure of the order of 300% at relatively high strain rates of approx 1 s exp —1. New results are described to evaluate the effect of silicon carbide whiskers on the high temperature deformation of the 2124 Al alloy. Recently, studies have demonstrated that some mechanically alloyed Al alloys can also exhibit superplastic characteristics with high elongations to failure of approx 500% at high strain rates of up to approx 10 s exp —1. The experimental results on these new Al-based alloys are described and reviewed with reference to superplasticity in conventional monolithic Al alloys. 30 ref.—AA

Category: 62 Controlled Terms: Aluminum base alloys/Composite materials / Silicon carbide / Composite materials/ Fiber composites/Mechanical properties/Superplasticity/Strain rate/Elongation. Alloys: 2124/AL

52-0221 METADEX Issue 8902
The Superplastic Forming of Sheet Metals.
Balbach, R.; Pross, K. Universitat Stuttgart Blech Rohre Profile May 1988; 35, (5); 335-337; ISSN 0006-4688; in German

Superplastic forming is exercised for various alloys, including Ti Ti6Al4V and Al Al6Zn2Mg1.5Cu, at approx 30% below the melting point of the metal. During superplastic forming, the material is elongated at a rate of 800-1000%. The most important requirement is a small particle size, with a 10 mu m maximum. The superplastic forming methods are primarily employed in the aerospace industry for the low volume production of special parts. 6 ref.—F.J.B.

Category: 52 Controlled Terms: Titanium base alloys/Forming/Aluminum base alloys/Forming/Airframes/Forming/ Superplastic forming/Microstructure. Alloys: Ti-6Al-4V/TI/Al6Zn2Mg1.5Cu/7475/AL

52-0219 METADEX Issue 8902
Production of Aerospace Parts Using Superplastic Forming and Diffusion Bonding of Titanium.
Anderson, T. T.; Hislop, L. Flameco Engineering. Superplasticity in Aerospace; Phoenix, Arizona, USA; 25-28 Jan. 1988; The Metallurgical Society/AIME; 420 Commonwealth Dr., Warrendale, Pennsylvania 15086, USA; 1988; 8902-72-0074; 345-360; in English

Superplastic forming (SPF), the most recent advancement in sheet metal fabrication technology, exploits the unique characteristics of the materials to undergo elongation of 750-1100%. Ti—6Al—4V exhibits this desired material behavior and is now being used to fabricate complex configurations not otherwise possible by conventional methods. Flameco Engineering, Inc. is currently engaged in the manufacturing of a number of production parts using the superplastic forming and diffusion bonding technology, using a variety of monolithic components, most of which are of one, two, three, or more sheet hollow structures. 7 ref.—AA

Category: 52 Controlled Terms: Titanium base alloys/Metal working/Aerospace / Superplastic forming/Diffusion welding/Aircraft components/Metal working. Alloys: Ti-6Al-4V/TI

52-0218 METADEX Issue 8902
Superplastic Forming of Inconel 718.
Mahoney, M. W.; Crooks, R. Rockwell International Science Center. Superplasticity in Aerospace; Phoenix, Arizona, USA; 25-28 Jan. 1988; The Metallurgical Society/AIME; 420 Commonwealth Dr., Warrendale, Pennsylvania 15086, USA; 1988; 8902-72-0074; 331-344; in English

There is interest in the use of high temperature alloys for reentry type aerospace vehicles and, concurrently, in the development of new manufacturing techniques to economically fabricate these alloys into usable shapes. Although Inconel 718 is not suitable for use in the most severely heated locations, it is a promising material candidate for potential fabrication of hot surface locations where temperatures remain < 700 deg C. The work presented illustrates that Inconel 718, specially fabricated as fine grain thin foil (< 0.5 mm thick), can be highly superplastic. Superplastic properties such as flow stress characteristics, strain rate sensitivity and total uniform elongation for both wrought and powder produced alloys are presented and discussed. Also, microstructural features such as grain size stability, cavitation and an evaluation of grain structures following superplastic forming are presented. For many properties, the wrought and powder alloys were very similar. However, the wrought alloy proved to be significantly superior to the powder alloy in total elongation (500 vs. 150%, respectively) and exhibited considerably less cavitation after superplastic forming. Evidence suggests that dynamic recrystallization occurs during superplastic forming, with flow behavior controlled by dislocation motion. 20 ref.—AA

Category: 52 Controlled Terms: Nickel base alloys/Metal working/Superalloys / Metal working/Superplastic forming/Spacecraft/Strain rate/Cavitation / Powder metallurgy parts/Metal working/Grain size. Alloys: Inconel 718/NI / SP

52-0216 METADEX Issue 8902
Advances in Superplastic Aluminum Forming.
Barnes, A. J. Superform USA Superplasticity in Aerospace; Phoenix, Arizona, USA; 25-28 Jan. 1988; The Metallurgical Society/AIME; 420 Commonwealth Dr., Warrendale, Pennsylvania 15086, USA; 1988; 8902-72-0074; 301-313; in English

The development of superplastic Al forming from its early beginning to its current status illustrates the increasing use of this technology by the aerospace industry. The emergence of newer alloys, such as Supral 220, SP7475, 8090 and 2090, heralds an expansion into more structural applications. The significance of process developments, such as back pressure forming and diffusion bonding, are discussed and the importance of designing for SPF is emphasized. 8 ref.—AA

Category: 52 Controlled Terms: Aluminum base alloys/Metal working / Superplastic forming/Aircraft components/Metal working/Bonding. Alloys: Supral 220/7475/8090/2090/AL

52-0212 METADEX Issue 8902
Cavitation in Aluminium Alloys During Superplastic Flow.
Pilling, J.; Ridley, N. Michigan Technological University, UMIST. Superplasticity in Aerospace; Phoenix, Arizona, USA; 25-28 Jan. 1988; The Metallurgical Society/AIME; 420 Commonwealth Dr., Warrendale, Pennsylvania 15086, USA; 1988; 8902-72-0074; 183-198; in English

Commercially important Al alloys (7475, Supral 220) are prone to cavitation during superplastic flow. Cavities either nucleate at grain boundary sites or pre-exist and their subsequent growth, coalescence and interlinkage, leads to premature failure. The presence of cavities in superplastically formed parts may have an adverse effect on their mechanical properties, while the existence of cavities in load bearing components, particularly for aerospace applications, is clearly undesirable. Several studies show that cavity growth is determinedprimarily by matrix plastic flow and that coalescence plays an important part in the development of large cavities. Hence, to prevent cavitation, it is necessary to inhibit the n nucleation event and to avoid the presence of pre-existing defects by care ful control of the processing required to develop the superplastic micro structure. The influence that microstructural features, such as grain size an d second phase particles, and deformation parameters, such as strain rate and tem perature, have on cavity nucleation is discussed. It is clear that it will be difficult to control the various factors so as to totally prevent cavitation damage during superplastic forming, without the use of additional remedial procedures. 35 ref.—AA

Category: 52 Controlled Terms: Aluminum base alloys/Metal working / Superplastic forming/Cavitation/Deformation effects/Grain size/Strain rate. Alloys: 7475/Supral 220/AL

52-0211 METADEX Issue 8902
Analysis of the Cone Test to Evaluate Superplastic Forming Characteristics of Sheet Metal.
Goforth, R. E.; Chandra, N. A.; George, D. Texas A&M University, Florida State University, LTV Aerospace. Superplasticity in Aerospace; Phoenix, Arizona, USA; 25-28 Jan. 1988; The Metallurgical Society/AIME; 420 Commonwealth Dr., Warrendale, Pennsylvania 15086, USA; 1988; 8902-72-0074; 149-166; in English

The laboratory cone-forming test was developed for the purpose of evaluating the superplastic forming characteristics of Ti alloys. The conical die was designed with a constant cone angle of 57-58 deg. In reality, the true strain rate varies throughout the test and only average values are obtained. This might be sufficient for comparing the relative spf characteristics of certain materials; however, for the purpose of establishing material parameters to be input into analytical process models, it leaves much to be desired. It should also be pointed out that Ti is much more tolerant to variations in strain-rate during the spf process than, for example, the Al spf alloys. A modified cone-test which will assure a constant true strain-rate throughout the complete test is proposed. During the first stage of the test, a pressure—time cycle is developed which will assure a constant strain rate. The die is designed with a varying cone angle, assuring a constant strain rate with a constant pressure throughout the second stage of the test. Materials tested using the modified cone test include: 2090 and 8090 Al—Li alloys. 12 ref.—AA

Category: 52 Controlled Terms: Titanium base alloys/Metal working/Aluminum base alloys/Metal working/ Superplastic forming/Testing/Strain rate. Alloys: 2090/8090/AL

52-0209 METADEX Issue 8902
The Forming Behaviour of Commercially Available Superplastic Aluminium Alloys.
Grimes, R.; Butler, R. G. British Alcan Aluminium, Superform Metals Superplasticity in Aerospace; Phoenix, Arizona, USA; 25-28 Jan. 1988; The Metallurgical Society/AIME; 420 Commonwealth Dr., Warrendale, Pennsylvania 15086, USA; 1988; 8902-72-0074; 97-113; in English

The forming behaviour of currently commercially available superplastic Al alloys is assessed in uniaxial and biaxial forming tests. Of the alloys assessed, those based on the original Al—6Cu—0.4Zr (Supral) system exhibited considerably greater superplastic capability over a wider temperature range and at higher strain rates than any of the other alloys. 7475E, however, was capable of achieving the highest strength and had a flow stress of about one-half (approx 5 MPa) that of the Supral alloys. From the point of view of structural aerospace use, the Al—Li based alloy, 8090, appears most attractive since its low quench sensitivity and high forming temperature allow simultaneous component manufacture and solution treatment without the need for water quenching. A number of Al—Mg alloys suitable for trim applications are also assessed. None of these possessed a superplastic capability matching the main Supral alloys. Nevertheless, the alloy Formall 548 gave encouraging results and, with somewhat improved superplastic capability, could be suitable for many trim components. 6 ref.—AA

Category: 52 Controlled Terms: Aluminum base alloys/Metal working / Superplastic forming/Formability/Ductility/ Temperature effects / Microstructure. Alloys: 8090/Supral 220/Supral 100/Supral 5000/7475 / Formall 700/Formall 545/Formall 548/AL

31-0630 METADEX Issue 8902
Memory Metal: Properties and Applications.
Besselink, P. A. Constructeur Aug. 1987; 26, (8); 28-37; ISSN 0010-6658; in Dutch

Memory alloys, which exhibit superplasticity, are finding increasing applications, especially in Japan. In the Netherlands, the interest has concentrated on the TiNi alloys, which have the best mechanical properties: Their structures are austenitic (high temperature) and martensitic (low temperature) the reversible transformation occurring with a hysteresis. The electrical properties change sharply during the transformation. The main applications are in medical technology, aerospace, robots, instruments, fasteners, energy conversion, sensors and actuators, and consumer goods. The compositions, properties, fabrications and future outlook of the alloys are discussed, with examples of current applications.—H.S

Category: 31 Controlled Terms: Titanium base alloys/Mechanical properties / Nickel/Alloying elements/Shape memory alloys/Mechanical properties / Martensitic transformations/Magnetic permeability/Thermal expansion / Resistivity/ Mechanical properties/Microstructural effects

31-0475 METADEX Issue 8902
Void Formation and Its Effect on Post-Formed Mechanical Properties in Superplastic 7475 Aluminum Alloy.
Eto, T.; Hirano, M.; Hino, M.; Miyagi, Y. Kobe Steel. Superplasticity in Aerospace; Phoenix, Arizona, USA; 25-28 Jan. 1988; The Metallurgical Society/AIME; 420 Commonwealth Dr., Warrendale, Pennsylvania 15086, USA; 1988; 8902-72-0074; 199-207; in English

Superplastic 7475 high strength Al alloy has been produced by commercial production root (approx 3 mm x 1100 w x 2200 L). It has a fine grain of approx 10 mu m in diameter and reveals › 500% in elongation, deformed at 516 deg C with the strain rate of 2 x 10 exp —4 s exp —1 . Using these superplastic 7475 Al alloy sheets, the cavitation and post-formed properties, the most important problems for their practical application, are discussed. The cavitation behavior was investigated in case of both uniaxial tension and biaxial forming such as gas pressure forming. The post-formed mechanical properties in T6, T76 and T73 conditions were investigated in both materials quenched in water and polymer quenchant to suppress distorsion during quenching. Cavitation started at the deformation of 50% in elongation and was affected by deformation mode. The cavitation in uniaxially tensile deformation was larger than in biaxial deformation. Voids were mainly formed at the triple point of grain boundaries, due to the decohesion through their sliding during deformation. Post-formed mechanical properties of deformed specimens in T6, T76 and T73 fully met those of the minimum value specified. They also have good corrosion resistance in T76 and T73 conditions. Related properties such as fatigue performance were good. The distortion in quenching operation was suppressed by polymer quenchant and the post-formed properties were almost equal to that quenched in water. The above results lead to the conclusion that superplastic 7475 Al alloy has promising potential for practical applications to aerospace. 9 ref.—AA

Category: 31 Controlled Terms: Aluminum base alloys/Mechanical properties / Superplasticity/Voids/Deformation effects/Yield strength/Exfoliation corrosion. Alloys: 7475/AL

8806-50305 Compendex
ADVANCED MATERIAL MAKES FIGHTER LIGHTER
Anon. Met Constr v 19 n 8 Aug 1987 p 444-445 Coden: MECOD ISSN: 0307-7896 In ENGLISH Refs: 7 refs. Doc. Type: JOURNAL ARTICLE Treatment Des.: Applications

Lital is one of a series of alloys, developed and patented by Alcan and the Royal Aircraft Establishment, which are around 10% lighter than conventional alloys of similar strength. Since these metals are also stiffer than conventional aerospace alloys they offer total savings of 15% or more in structure weight in aircraft designed to take full advantage of their properties. The alloy has recently been specified for the structural parts of a European fighter aircraft

Card-A-Lert Codes: 415 (Metals, Plastics, Wood and Other Structural Materials) / 662 (Automobiles and Smaller Vehicles)/541 (Aluminum and Alloys)/542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/549 (Nonferrous Metals and Alloys in General)/404 (Civil Defense and Military Engineering)/Controlled Terms: (*AIRCRAFT MATERIALS — *Light Metals) / (ALUMINUM LITHIUM ALLOYS — Applications)/(AIRCRAFT — MILITARY — Fighter) / (METALS AND ALLOYS — Superplasticity)/Uncontrolled Terms: ADVANCED LIGHTWEIGHT ALLOYS/ EUROPEAN FIGHTER AIRCRAFT

8806-50368 Compendex
VERSATILE METAL-MATRIX COMPOSITES
Niskanen, P.; Mohn, W. R. Advanced Composite Materials Corp, Greer, SC, USA Adv Mater Processes v 133 n 3 Mar 1988 p 39-41 Coden: AMAPE ISSN: 0882-7958 In ENGLISH. Doc. Type: JOURNAL ARTICLE Treatment Des.: Applications GENERAL REVIEW

Discontinuous silicon carbide (SiC) reinforced aluminum metal-matrix composites (MMC) are as light as aluminum but have higher strength and specific stiffness. Some grades are isotropic and have better compressive microcreep resistance than beryllium. In addition, they can be tailored to match the coefficients of thermal expansion of other materials including beryllium, stainless steel, and electroless nickel. This class of advanced engineering materials which can be forged, superplastically formed, and precision machined into complex shapes, has recently been qualified for use in aerospace structures, inertial guidance systems, and lightweight optical assemblies

Card-A-Lert Codes: 541 (Aluminum and Alloys)/415 (Metals, Plastics, Wood and Other Structural Materials)/652 (Aircraft)/542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/Controlled Terms: (*ALUMINUM AND ALLOYS — *Metallic Matrix Composites)/(AIRCRAFT MATERIALS — Light Metals) / Uncontrolled Terms: DISCONTINUOUS SILICON CARBIDE REINFORCEMENTS/MECHANICAL PROPERTIES/FRACTURE TOUGHNESS/MODULUS OF ELASTICITY/SILICON CARBIDE WHISKER REINFORCEMENTS/ DISCONTINUOUS SILICON CARBIDE REINFORCEMENT

8805-40892 Compendex
ENGINEERED METAL MATRIX COMPOSITES FOR PRECISION OPTICAL SYSTEMS
Mohn, Walter R.; Vukobratovich, Daniel. Advanced Composite Materials Corp, Greer, SC, USA. SAMPE J v 24 n 1 Jan-Feb 1988 p 26-35 Coden: SAJUA ISSN: 0036-0813 In ENGLISH Refs: 13 (Author abstract) 13 refs; Doc. Type: JOURNAL ARTICLE Treatment Des.: Applications GENERAL REVIEW EXPERIMENTAL

Discontinuous silicon carbide reinforced aluminum metal matrix composites (MMC) are a unique class of advanced engineered materials which have been developed and recently qualified for use in aerospace structures, inertial guidance systems, and lightweight optical assemblies. Such materials are as light as aluminum but exhibit significantly greater strengths and specific stiffness. Certain grades of these MMC's are isotropic and are more resistant to compressive microcreep than beryllium; and they can be tailored to match the coefficients of thermal expansion (CTE) of other materials, including beryllium, stainless steel, and electroless nickel. Since these composites can be easily forged, superplastically formed, and precision machined into complex shapes, they are ideal for use in the economical production of stable optical systems. This paper describes some of the enhanced properties of engineered MMC's, discusses some of the design considerations that have led to the specification of these materials for building an ultralightweight telescope, and presents some interesting results obtained from prototype testing

Card-A-Lert Codes: 541 (Aluminum and Alloys)/941 (Acoustical and Optical Measuring Instruments)/531 (Metallurgy and Metallography)/741 (Light, Optics and Optical Devices)/812 (Ceramics and Refractories)/Controlled Terms: (*ALUMINUM AND ALLOYS — *Metallic Matrix Composites)/(OPTICAL INSTRUMENTS — Materials)/ (TELESCOPES — Materials)/SILICON CARBIDE/(METALS AND ALLOYS — Fiber Reinforcement)/MIRRORS/Uncontrolled Terms: METAL MATRIX COMPOSITES / PRECISION OPTICAL SYSTEMS

88A45201 NASA IAA Meeting Paper Issue 18
Aluminum-lithium alloys: Design, development and application update; Proceedings of the Symposium, Los Angeles, CA, Mar. 25, 26, 1987 (Book) (AA)KAR, RAMESH J.; (AB)AGRAWAL, SUPHAL P.; (AC)QUIST, WILLIAM E. (AA)ED.; (AB)ED.; (AC)ED. (AB)(Northrop Corp., Aircraft Div., Hawthorne, CA); (AC)(Boeing Commercial Airplane Co., Seattle, WA). Symposium organized and sponsored by ASM International. Metals Park, OH, ASM International, 1988, 470 p. For individual items see A88-45202 to A88-45205. 880000 p. 470 In: EN (English) Members, $62.40; nonmembers, $78 p.3030

The present conference on the development status of aluminum-lithium alloys for aerospace applications discussed topics in the availability of these alloys, their fatigue, fracture, and corrosion characteristics, their design criteria, and manufacturing techniques developed for them to date. Attention is given to developments in rapidly-solidified Al-Li alloys, the mechanisms of fatigue crack propagation in commercial Al-Li alloys, the effects of processing on Al-Li microstructures and fracture behavior, and Al-Li exfoliation and stress corrosion cracking behavior. Also discussed are design considerations for novel aerospace vehicle materials, critical Al-Li alloy design factors, the äapplication of Al-Li alloys in naval aircraft, and the superplastic forming characteristics of Al-Li sheet alloys. O.C.

Category code: 26 (metallic materials) Controlled terms: *ALUMINUM ALLOYS /*CONFERENCES /*LITHIUM ALLOYS /*MECHANICAL PROPERTIES/AIRCRAFT CONSTRUCTION MATERIALS/CRACK PROPAGATION/ F-15 AIRCRAFT/FATIGUE TESTS/FORMING TECHNIQUES/INVESTMENT CASTING/METAL SHEETS/SPACECRAFT CONSTRUCTION MATERIALS/STRESS CORROSION CRACKING/STRUCTURAL DESIGN/SUPERPLASTICITY / TECHNOLOGY ASSESSMENT /

88A43932# NASA IAA Conference Paper Issue 18
Superplasticity in structural Al alloys
(AA)GHOSH, A. K. (AA)(Rockwell International Science Center, Thousand Oaks, CA) F33615-79-C-3218; F49620-83-C-0055 IN: Interdisciplinary issues in materials processing and manufacturing; Proceedings of the Symposium, ASME Winter Annual Meeting, Boston, MA, Dec. 13-18, 1987. Volume 2 (A88-43926 18-37). New York, American Society of Mechanical Engineers, 1987, p. 437-449. Research supported by Rockwell Independent Research and Development Programs. 870000 p. 13 refs 20 In: EN (English) p.3028

A brief review of superplasticity in a number of Al alloys is presented. Emphasis is placed on an Al-Zn-Mg-Cu alloy (7475), which possesses high structural strength suitable for aerospace applications. Discussion is also presented on the superplasticity in Al-Li alloys. Recent developments in thermomechanical processing have led to significant superplasticity in these materials. Some new aspects of microstructural changes and mechanical behavior of these materials are elucidated. A short discussion on cavitation and its control is also presented. Author

Category code: 26 (metallic materials) Controlled terms: *ALUMINUM ALLOYS /*MECHANICAL PROPERTIES / *MICROSTRUCTURE /* SUPERPLASTICITY /*THERMOMECHANICAL TREATMENT/COPPER ALLOYS/CRYSTAL STRUCTURE/GRAIN SIZE/LITHIUM ALLOYS / MAGNESIUM ALLOYS/ZINC ALLOYS /

62-0655 METADEX Issue 8812
Properties of High-Strength Superplastic PM and PM/MMC Aluminum Alloys.
Mahoney, M. W.; Kendig, M.; Murphy, A. R.; Mitchell, M. R.; Ghosh, A. K. Rockwell International Science Center. PM Aerospace Materials '87; Luzern, Switzerland; 2-4 Nov. 1987; MPR Publishing Services Ltd.; Old Bank Buildings, Bellstone, Shrewsbury SY1 1HU, UK; 1988; 8812-72-0634; 33.1-33.14; in English

Recent results have shown that a high-strength PM Al alloy, 7064, and this same alloy reinforced with SiC sub p, can be thermomechanically processed to produce a fine equiaxed grain size. With this microstructure, high levels of superplasticity are achievable and, accordingly, complex-shaped aerospace parts were superplastically formed. A brief review of this work is presented, highlighting PM processing, superplastic characteristics and part forming. Following this synopsis, a detailed presentation is given, illustrating additional properties of both the PM-7064 Al alloy and the PM-7064 Al/SiC sub p metal matrix composite alloy. Pitting corrosion results of both the powder alloy and reinforced alloys, after fine-grain processing, illustrate pitting potential for solution-treated and full-strength aged conditions. Also, fatigue crack growth rate is presented with an analysis of the fatigue crack path based on microstructural details. 29 ref.—AA

Category: 62. Controlled Terms: Aluminum base alloys/Composite materials / Silicon carbide / Composite materials/ Composite materials/Powder technology / Powder metallurgy parts/Corrosion/Superplasticity/Pitting (corrosion) / Sodium chloride/Environment. Alloys: 7064/AL

61-0746 METADEX Issue 8812
Superplastic Forming/Weld-Brazing of Ti—6Al—4V Panels for Enhanced Structural Efficiency.
Royster, D. M.; Bales, T. T.; Davis, R. C. NASA Langley Research Center. Competitive Advances in Metals and Processes—1st International SAMPE Metals and Metals Processing Conference. Vol. 1; Cherry Hill, New Jersey, USA; 18-20 Aug. 1987; Society for the Advancement of Material and Process Engineering; P.O. Box 2459, Covina, California 91722, USA; 1987; 8812-72-0696; 191-202; in English

The objectives of these NASA studies are to exploit the processing advantages of superplastic forming and joining by weld-brazing (WB) for the fabrication of Ti skin-stiffened structural components. Application of these advanced stiffener shapes offers the potential to achieve substantial weight savings in aerospace vehicles. 6 ref.—AA

Category: 61. Controlled Terms: Titanium base alloys/Fabrication/Panels / Fabrication/Superplastic forming/Spot welding/Brazing/Weight reduction

8804-30562 Compendex
DESIGN AND MANUFACTURE OF A SUPERPLASTIC-FORMED ALUMINUM-LITHIUM COMPONENT
Henshall, C. A.; Wadsworth, J.; Reynolds, M. J.; Barnes, A. J. Lockheed Missiles & Space Co. Palto Alto, CA, USA. Mater Des v 8 n 6 Nov-Dec 1987 p 324-330 Coden: MADSD ISSN: 0264-1275 In ENGLISH Refs: fs. Doc. Type: JOURNAL ARTICLE Treatment Des.: GENERAL REVIEW

The phenomenon of superplasticity in metallic alloys is briefly described. The design, manufacture and post-forming analyses are described of a complex shape produced by superplastic forming using the aluminum-lithium alloy 2090. Complex shapes for a wide variety of aerospace and commercial applications have been successfully formed. An example of a component superplastically formed with SUPRAL is shown. (Edited author abstract) 14

Card-A-Lert Codes: 541 (Aluminum and Alloys)/542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/549 (Nonferrous Metals and Alloys in General)/535 (Rolling, Forging and Forming)/421 (Strength of Materials; Mechanical Properties)/Controlled Terms: (*ALUMINUM LITHIUM ALLOYS — *Superplasticity)/(METALS TESTING — Tensile Tests)/(METAL FORMING — Strain)/Uncontrolled Terms: ALUMINUM LITHIUM/SUPERPLASTICITY/ COMPLEX SHAPE

88A40485 NASA IAA Journal Article Issue 16
Aerospace materials for the 21st century
(AA)CANNON, PETER (AA)(Rockwell International Corp., El Segundo, CA). Journal of Metals (ISSN 0148-6608), vol. 40, May 1988, p. 10-14. 880500 p. 5 In: EN (English) p.2611

A development status evaluation and prospective performance characteristics projection is made for advanced metal-matrix composites, ceramic-matrix composites, high-temperature polymeric composite resins, and the NDE and process-monitoring techniques that may be employed in order to maximize their performance while minimizing their costs. Attention is given to P/M aluminum alloy matrix-based, superplastically formed SiC whisker-reinforced composites, and the materials requirements of the National Aerospace Plane and NASA Space Station. O.C.

Category code: 23 (chemistry/materials) Controlled terms: *AEROSPACE INDUSTRY /*ALUMINUM ALLOYS / *CERMETS /*LOW DENSITY MATERIALS /*MAGNESIUM ALLOYS /*METAL MATRIX COMPOSITES/GRAIN SIZE/HIGH STRENGTH ALLOYS/SILICON CARBIDES / SUPERPLASTICITY /

46-0379 METADEX Issue 8812
Solid State Microblend Microcomposite Materials for Defense Applications.
Patel, A. N.; Diamond, S.; Erich, D.; Goddard, S. Battelle Columbus. Powder Metallurgy in Defense Technology. Vol. 7; Annapolis, Maryland, USA; 3-4 Dec. 1986; Metal Powder Industries Federation; 105 College Rd. East, Princeton, New Jersey 08540, USA; 1987; 8812-72-0675; 159-174; in English

The demand by industry and the military for new alloys and improved materials for use in both conventional and special environments has prompted vigorous research and development activities in all phases of materials processing and, more recently, a revived interest in alloy development. In addition, the economic and strategic necessity to conserve energy and resources has prompted further interest in the development of new alloys. A number of the new alloys have been prepared both by rapid solidification and by solid state microblending (SSMB). Similar unique properties and unconventional microstructures have been achieved by both powder methods. Only the results of solid state processing, however, are described. Battelle's columbus Division (BCD) has been actively participating in these research areas and has made significant contributions to the application and extension of solid state microblending to produce the desired powders and to the subsequent consolidation of the powders into bulk forms. The developments of these technologies have increased understanding of their use to produce alloys having unique properties. Some of these alloys are now being introduced into commercial applications requiring improved performance and are also being developed for wholly new applications. The new microstructures and considerably altered second phase distribution and morphology in these materials of non-conventional chemistry contrast sharply with those of their conventionally processed counterparts. Based on these unique microstructures and the ability to control the material properties, such systems are expected to have wide applications as structural materials, lightweight alloys for automotive, aircraft, and aerospace use, high-temperature alloys, fine particle permanent magnets, electrical contacts and batteries, bearings, and superplastic materials. Undoubtedly, other applications will emerge as these development techniques evolve further. 9 ref.—AA

Category: 46. Controlled Terms: Copper base alloys/Alloydevelopment/Lead base alloys/Alloy development/Nickel base alloys/Alloy development / Titanium base alloys/Alloy development/Ferrous alloys/Alloy development / Metal powders/Alloy development/Mechanical alloying/Aircraft components / Materials selection/Automotive components/ Materials selection/Permanent magnets/Materials selection/Electric batteries/Materials selection

61-0687 METADEX Issue 8811
Materials Trends in Military Airframes.
Stubbington, C. A. Royal Aerospace Establishment. Met. Mater. July 1988; 4, (7); 424; ISSN 0266-7185; in English

The dominant position held by aluminium alloys in airframe construction has been challenged by composite materials. The materials and processing techniques currently available to designers of military aircraft are reviewed and some of the developments which are likely to influence the choice of materials and processes for future projects are considered. 35 ref.—AA

Category: 61 Controlled Terms: Airframes/Materials selection/Aluminum base alloys/Mechanical properties/ Composite materials/Mechanical properties / Titanium base alloys/Mechanical properties/High strength steels/Mechanical properties/Superplastic forming/Diffusion welding. Alloys: 2024/2324/ 7475/8090/8091/7050/7178/7075/2014/7010/ AL/Ti-5Al-2.5Sn/Ti-6Al-4V/Ti-4Al-4Mo-2Sn-0.5Si/Ti-4Al-4Mo-0.4Sn-0.5Si/TI/AZ91/WE54/MG/35NCD16/ SANCM

88A36913 NASA IAA Conference Paper Issue 14
Superplasticity in SiC reinforced Al alloys
(AA)MAHONEY, M. W.; (AB)GHOSH, A. K.; (AC)BAMPTON, C. C. (AB)(Rockwell International Science Center, Thousand Oaks, CA); (AC)(North American Aircraft Operations, El Segundo, CA). F33615-83-C-3235 IN: International Conference on Composite Materials, 6th, and European Conference on Composite Materials, 2nd, London, England, July 20-24, 1987, Proceedings. Volume 2 (A88-36851 14-24). London and New York, Elsevier Applied Science, 1987, p. 2.372-2.381. Research supported by the Rockwell International Independent Research and Development Fund. 870000 p. 10. refs 14 In: EN (English) p.2238

Recent studies and results on SiC reinforced aluminum alloys show that significant benefits in properties can be achieved over base alloy compositions. However, because of their increase in modulus or stiffness, metal-matrix composite (MMC) alloys are difficult to form into usable shapes by conventional sheet forming practices. This inability to fabricate parts limits the implementation of this new class of material. With appropriate thermomechanical processing, it is possible to achieve a fine grain size in SiC-reinforced aluminum alloys and accordingly achieve sufficient superplasticity to gas pressure-form complex shaped aerospace parts. Author

Category code: 24 (composite materials) Controlled terms: *ALUMINUM ALLOYS /*MECHANICAL PROPERTIES /*METAL MATRIX COMPOSITES /*SILICON CARBIDES /*SUPERPLASTICITY /THERMO-MECHANICAL TREATMENT/GRAIN SIZE/LITHIUM ALLOYS/MICROSTRUCTURE/SPACECRAFT CONSTRUCTION MATERIALS /

88A35113 NASA IAA Conference Paper Issue 13
Advanced aluminum alloys and aluminum-polymer laminate for aerospace structure
(AA)STALEY, JAMES T. (AA)(Alcoa Laboratories, Alcoa Center, PA). AAS PAPER 86-381 IN: Aerospace century XXI: Space flight technologies; Proceedings of the Thirty-third Annual AAS International Conference, Boulder, CO, Oct. 26-29, 1986 (A88-35093 13-12). San Diego, CA, Univelt, Inc., 1987, p. 1001-1010. 870000 p. 10 refs 6 In: EN (English) p.2029

Advanced Al alloys and hybrid aluminum-polymer laminates furnish highly cost-effective competition not only to established Al and Ti alloys, but to graphite/epoxy and metal-matrix composites, in aerospace structure applications. The new Al-Li alloys are more tractable in design due to their isotropic nature, and involve substantially lower costs. Hybrid laminates have 10-20 percent lower density than monolithic conventional Al alloy sheet, as well as excellent vibration damping, and may prove ideal in fatigue-critical applications. O.C.

Category code: 23 (chemistry/materials). Controlled terms: *ALUMINUM ALLOYS /*LAMINATES /*POLYMERS /*SPACECRAFT STRUCTURES/CERIUM/LITHIUM ALLOYS / POLYAMIDE RESINS/SUPERPLASTICITY / WROUGHT ALLOYS /

NTN88-0122/XAD NTIS
Superplastically Formed Aluminum-Base Alloys
Department of the Air Force, Washington, DC (000260000) NTIS Tech Note, Mar 88, 1p, FOR ADDITIONAL INFORMATION: Detailed information about the technology described may be obtained by ordering the NTIS report order number AD-A161366/0/NAC, price code A06, NTIS Prices: See availability statement

This citation summarizes a one-page announcement of technology available for utilization. Properties of three alloys designed specifically for aerospace application in the superplastically formed condition, Supral 100, Supral 150 clad, and Supral 220 clad, were determined in studies conducted at Fairchild Republic Company for the Air Force. Tests were conducted on superplastically formed parts in the T6 temper under conditions providing MIL-HDBK-5 data. The results of these tests are given both in tabular and graphic form. Several trends were observed that were true for all three alloys. It was noted, for example, that properties were approximately the same whether determined parallel to or perpendicular to the direction of superplastic strain. Also, in all three alloys the amount of superplastic forming did not appear to be an important factor unless it was sufficiently high. The Supral alloys were reported to be comparable to 7475 superplastic alloy in overall formability. However, forming of supral alloys can be accomplished at higher strain rates and lower forming temperature, factors which appreciably reduce forming costs

Fld: 71N Controlled terms: *Aluminum alloys/*Aerospace engineering /Uncontrolled terms: NTISNTND

52-2128 METADEX Issue 8810
Recent Research and Development of Superplasticity.
Miyagawa, M.; Kobayashi, M. Bull. Jpn. Inst. Met. 1986; 25, (1); 8-15; ISSN 0021-4426; in Japanese

Japanese research into superplastic forming coupled with diffusion bonding has centred largely on alloys, such as Zn—22Al and Ni and Ti base systems. Particular applications are found in aerospace engineering. The advantages of superplastic forming include the saving of process time and materials by near net shaping. Computer aided design can be applied to the process. A major requirement of superplastic materials is resistance to crack initiation and propagation. 21 ref.—J.G.

Category: 52 Controlled Terms: Zinc base alloys/Metal working/Aluminum / Alloying elements/Superplastic forming/Diffusion welding/Crack propagation Alloys: Zn-22Al/ZN

52-2049 METADEX Issue 8810
Superplastic Forming for Tomorrow's Metal Manufacturing.
East, W. R. Mater. Eng. (Cleveland) Apr. 1988; 105, (4); 37-40; ISSN 0025-5319; in English

Superplastic forming (SPF) began with experiments with Bi—Sn alloys in the 1930s, but remained essentially a laboratory curiosity until the 1960s when other alloys were found to exhibit superplastic properties. A Zn—Al alloy was used to produce the first commercial SPF products. Today, most of the SPF is for Ti, Al and Al—Li alloys for the aerospace industry, together with some work on superplastic forming stainless steel components. A SPF alloy must have a very fine grain size (< approx 8 mu m) and be equiaxed. The grain size must remain small at temperatures (= 90% of the material's melting temperature for 2 h or longer. Alloys meeting these specifications include: Al alloys 2004, 2419 and thermomechanically processed 7475; Al—Li alloys such as 2090, 2091 and 8090; most of the duplex stainless steels with an equal balance of austenite and ferrite, such as 2205 series; and all of the alpha — beta Ti alloys such as Ti—6Al—4V.—D.O.N.

Category: 52 Controlled Terms: Aluminum/Metal working/Titanium/Metal working/Aluminum base alloys/Metal working/Duplex stainless steels/Metal working/Titanium base alloys/Metal working/Superplastic forming/Strain rate/Aircraft/Materials selection/Aircraft components/Materials selection. Alloys: 2004/2419/7475/2090/2091/8090/AL/2205/SSD / Ti-6Al-4V/TI

31-3669 METADEX Issue 8810
On the Effects of Hydrostatic Pressure on Mechanical Properties of 7475 Aluminum Alloy.
Franklin, J. E.; Mukhopadhyay, J.; Mukherjee, A. K. University of California, Aerojet TechSystems. Scr. Metall. June 1988; 22, (6); 865-870; ISSN 0036-9748; in English

Superplasticity is now a viable near-net-shape forming process with considerable potential for industrial application. One particular Al alloy, e.g. 7475 Al, has received special attention in this regard for aerospace application. However, there is a practical limitation for using 7475 alloys commercially, because they form intergranular cavities. These cavities form during superplastic deformation and they limit the ductility and degrade the mechanical properties of post formed components. Hence, significant attention has been paid by investigators in recent years to minimize this cavitation phenomenon. Several investigators have used hydrostatic gas pressure as the main tool for reducing cavities in superplasticity. Such gas pressure had previously been used for reducing cavities in high temperature creep deformation. The mechanical behavior and microstructural aspect of 7475 Al alloys under a uniaxial test condition and the effect of hydrostatic gas pressure to a pressure level higher than that investigated earlier were studied. 15 ref.—AA

Category: 31 Controlled Terms: Aluminum base alloys/Mechanical properties / Superplasticity/Pressure effects/Cavitation/Pressure effects/Fracture strength/Pressure effects/Hydrostatic pressure. Alloys: 7475/AL

88A29724 NASA IAA Journal Article Issue 11 **Metals or plastics? MBB studies materials for future lightweight engineering**
New-Tech News, no. 1, 1988, p. 24-27. 880000 p. 4 In: EN (English) p.0

High-quality, high-strength materials and production methods for complex lightweight structures are nearly always associated with comparatively high costs, so that clear cost criteria must be established for their use. Attention is presently given to the comparative advantages of polymeric-matrix composites and metal-matrix composites, superplastically formable/diffusion weldable alloys of Al and Ti, and P/M metallic materials in aerospace applications. O.C.

Category code: 23 (chemistry/materials) Controlled terms: *AIRCRAFT CONSTRUCTION MATERIALS / *MATERIALS SCIENCE /*METAL MATRIX COMPOSITES /*POLYMER MATRIX COMPOSITES /*REINFORCED PLASTICS /*WEIGHT REDUCTION/AIRCRAFT PARTS/ALUMINUM ALLOYS/FIBER COMPOSITES/TITANIUMALLOYS /

62-0338 METADEX Issue 8807
Superplasticity in SiC Reinforced Aluminum Alloys.
Mahoney, M. W.; Ghosh, A. K.; Bampton, C. C. Rockwell International, North American Aircraft Operations. Sixth International Conference on Composite Materials and Second European Conference on Composite Materials (ICCM & ECCM). Vol. 2; London, UK; 20-24 July 1987; Elsevier Applied Science Publishers Ltd.; Crown House, Linton Rd., Barking, Essex IG11 8JU, UK; 1987; 8807-72-0356; 2.372-2.381; in English

Recent studies and results on SiC reinforced Al alloys show that significant benefits in properties can be achieved over base alloy compositions. However, because of their increase in modulus or stiffness, metal-matrix composite (MMC) alloys are difficult to form into usable shapes by conventional sheet forming practices. This inability to fabricate parts limits the implementation of this new class of material. With appropriate thermomechanical processing, it is possible to achieve a fine grain size in SiC-reinforced Al alloys and accordingly achieve sufficient superplasticity to gas pressure form complex shaped aerospace parts. 14 ref.—AA

Category: 62 Controlled Terms: Silicon carbide/Composite materials/Aluminum base alloys / Composite materials/ Whisker composites/Superplasticity / Grain size. Alloys: 7475/7075/7091/AL

52-0847 METADEX Issue 8805
Superplastic Forming of Titanium Alloy Sheets.
Inoue, M.; Takahashi, A.; Tsuzuku, T. Sumitomo Light Metals. Sumitomo Light Met. Tech. Rep. Oct. 1987; 28, (4); 50-57; ISSN 0039-4963; in Japanese

T—6Al—4V alloy has been used most extensively among Ti alloys in aerospace industries. The alloy sheet is usually formed by hot creep forming because of its small elongation and large spring back at room temperature. The alloy, however, exhibits prominent superplasticity and diffusion weldability with suitable conditions. Some studies on the superplastic forming and the combined process of superplastic forming/diffusion bonding were conducted for the development of integrated Ti alloy sheet structures in airframes for cost savings. The superplasticity, the superplastic forming, the superplastic forming/ diffusion bonding and some applications are described. 6 ref.—AA

Category: 52 Controlled Terms: Titanium base alloys/Forming/Superplastic forming/Diffusion welding. Alloys: Ti-6Al-4V/TI

31-1566 METADEX Issue 8805
Development and Properties of Superplastic Aluminum Alloys.
Matsuki, K. Bull. Jpn. Inst. Met. 1987; 26, (4); 263-271; ISSN 0021-4426; in Japanese

Topics about superplasticity of Al-alloys are reviewed. One of the most important incentives for development of superplasticity is its application to instruments and parts of aerospace industries. Deformation process by grain boundary slip and main alloys, i.e. 7000 group, Al—Li alloys, and Al—SiC whisker composites, grain refinement, cavity formation and cavity suppression are explained. Basic researches and new alloy developments intended for practical applications are reviewed. 104 ref.—JLMA

Category: 31 Controlled Terms: Aluminum base alloys/Mechanical properties / Lithium/Alloying elements/ Superplasticity/Silicon carbide/Composite materials/Composite materials/Mechanical properties. Alloys: 2124/7075 / 7475/8090/Al-6Cu-0.4Zr/Supral 100/Al-6Cu-0.4Zr-0.2Mg-0.1Ge/Supral 220 / Al-5Zn-5Co/8050/AL/Al-5Mg-0.6Cu-0.7Mn-0.15Cr/Al-2Li-3Cu-1Mg-0.2Zr/AL

88A13198 NASA IAA Conference Paper Issue 03
Development of particulate reinforced high strength aluminium alloy for aerospace applications
(AA)KRISHNADAS NAIR, C. G.; (AB)DUTTA, D.; (AC)KRISHNADEV, M. R. (AB)(Hindustan Aeronautics, Ltd., Bangalore, India); (AC)(Universite Laval, Quebec, Canada). IN: Advanced materials technology '87; Proceedings of the Thirty-second International SAMPE Symposium and Exhibition, Anaheim, CA, Apr. 6-9, 1987 (A88-13126 03-23). Covina, CA, Society for the Advancement of Material and Process Engineering, 1987, p. 889-901. 870000 p. 13 In: EN (English) p.295

A SiC particulate-reinforced, high strength Zr-refined Al-Zn-Mg-Cu alloy has been produced by PM. Powder prepared by gas atomization is mixed with SiC particulates in vacuum, followed by hot pressing and extrusion. Microstructure and properties of the composite are discussed. The resulting alloy is forgeable and heat-treatable to develop high strength. The alloy is economical to produce and is considerably cheaper than the Al-Li alloys. Author

Category code: 24 (composite materials) Controlled terms: *AIRFRAMES /*ALUMINUM ALLOYS /*METAL MATRIX COMPOSITES /* SILICON CARBIDES/CHEMICAL COMPOSITION / LITHIUM ALLOYS/ MECHANICAL PROPERTIES/SUPERPLASTICITY/TENSILE PROPERTIES /

52-0415 METADEX Issue 8803 **Analyses of Axisymmetric Sheet Forming Processes by Rigid-Viscoplastic Finite Element Method.** Park, J. J.; Oh, S. -I.; Altan, T. Battelle Columbus, Ohio State University J. Eng. Ind. (Trans. ASME) Nov. 1987; 109, (4); 347-354; ISSN 0022-0817; in English

Two types of sheet forming processes are analyzed by rigid-viscoplastic FEM (finite element method): axisymmetric punch stretching and hydrostatic bulge forming. The present formulations, based on the membrane theory and the Hill's anisotropic flow rule, include the rate sensitivity which is a key factor in controlling the forming of superplastic materials. Normal anisotropy is taken into account and Coulomb friction is assumed at the interface between punch and sheet. Nonsteady-state deformation processes, investigated in this study, were quasi-statically and incrementally analyzed. An FEM code was developed, using two-node linear elements with two degrees of freedom at each node, and applied to solve four categories of problems: A.K. steel punch stretching; hydrostatic bulging of a rate-insensitive material (Ti—6Al—4V, Al—

6Cu—0.4Zn, Zn—22Al); hydrostatic bulging of rate-sensitive materials; and hydrostatic bulging of a superplastic material (Ti-6-4). Strain distributions and shape changes predicted in the first two problems were compared with experiments and results of other analyses. The results of the third problem could not be compared with experiments; however, the results showed that the rate sensitivity affects the deformation as expected. The fourth problem is the main theme of the paper. To maintain the superplasticity in forming processes and to produce sound products, the control of the strain-rate is a key factor. A hydrostatic bulge forming process, which is often used for manufacturing structural aerospace parts, was analyzed and discussed. Further, an optimum pressure curve (pressure vs. time), which maintains the desired strain-rate in the deformed material, was obtained and compared with the results of an analytical prediction. 24 ref.—AA

Category: 52. Controlled Terms: Titanium base alloys/Forming/Stretch forming / Hydrostatic forming/Superplastic forming/Mathematical models/Aluminum base alloys/Forming/Zinc base alloys/Forming/Steels/Forming. Alloys: Ti-6Al-4V/TI/Al-6Cu-0.4Zn/AL/Zn-22Al/ZN

61-0151 METADEX Issue 8802
Designing for Superplastic Alloys.
Stephen, D. British Aerospace. AGARD (NATO) Aug. 1987; 7.1-7.37; ISSN 0365-2475; in English

Twenty years of development have brought the processes of superplastic forming and diffusion bonding to a state of maturity. These processes provide the Project Designer with the opportunity to design components on new projects which are both cost and weight efficient. However, to achieve optimum performance, the designer needs to have an in-depth understanding of the freedoms and limitations provided by these processes. Substantial evidence exists to support the claim that titanium alloys, when processed by the SPF/DB route, can compete in weight and more particularly cost, with conventional aluminium fabrication. This is likely to be a major factor in the future exploitation of these processes and clearly requires a revision to the designers traditional views of the areas of application for titanium alloys. The more recent developments in the processing of high strength superplastic aluminium alloys will clearly add to the further use of the superplastic forming process on future aerospace products, but the development of a combined SPF/DB process for aluminium alloys, with the full range of capabilities provided currently by titanium alloys, remains to be established. 33 ref.—AA

Category: 61 Controlled Terms: Aluminum base alloys/Alloy development / Titanium base alloys/Alloy development/ Diffusion welding/Superplastic forming/Welded joints/Mechanical properties/Aircraft components / Fabrication/Weight reduction/Design

55-0557 METADEX Issue 8802
Diffusion Bonding of Metals.
Partridge, P. G. AGARD (NATO) Aug. 1987; 5.1-5.23; ISSN 0365-2475; in English

The need to reduce the cost and weight of aerospace metallic structures has led to increased interest in solid state and liquid phase diffusion bonding processes especially in combination with superplastic forming. The bonding mechanisms and bonding techniques are reviewed and the process variables that affect bond quality and strength are described with reference to bonds between Ti-alloys, Al-alloys and dissimilar metals. The importance of quality control and the limitations of current NDE techniques for diffusion bonding are emphasised. Finally some trends and priorities in diffusion bonding technology are indicated. 74 ref.—AA

Category: 55 Controlled Terms: Titanium base alloys/Joining/Aluminum base alloys/Joining/Dissimilar metals/Joining/ Diffusion welding/Welded joints/Mechanical properties/Nondestructive testing/Quality control / Reviews. Alloys: Ti-6Al-4V/TI/7475/7010/AL

52-0232 METADEX Issue 8802
Superplastic Forming Comes of Age.
Peck, D. Mach. Tool Blue Book Nov. 1987; 82, (11); 50-54; ISSN 0024-9106; in English

Superplastic forming (SPF) of Al alloy sheet is applicable where relatively high-strength/low-weight properties are required in complex formed parts, but it is not competitive with more conventional metal-forming or plastic-molding methods for high-volume, low-cost components. The four basic types of SPF dies are female, female drape form, male, and male drape form. The primary advantages of SPF are that the tooling is simple and relatively inexpensive. SPF is used extensively in the US for aircraft and aerospace components.—A.R.

Category: 52 Controlled Terms: Aluminum base alloys/Forming/Superplastic forming/Aircraft components/Forming. Alloys: 2004/Supral 100/Supral 220 / AL

31-0682 METADEX Issue 8802
The Mechanical Properties of Superplastically Formed Titanium and Aluminium Alloys
Partridge, P. G.; McDarmaid, D. S. Royal Aircraft Establishment, British Aerospace. AGARD (NATO) Aug. 1987; 6.1-6.23; ISSN 0365-2475; in English

The behaviour of aluminium and titanium alloys under superplastic forming conditions is well documented but there is much less published data on the effect of the superplastic forming process on the mechanical properties; these data are essential for the design of structures. This paper reviews the effect of superplastic forming parameters such as temperature, strain rate, strain and post forming heat treatments upon the tensile, fatigue and fatigue crack growth performance of these alloys and relates the property variation to changes in the microstructure. During superplastic forming of aluminium alloys intergranular cavities are formed with increasing strain which degrade the material and reduce the mechanical properties. Ways to prevent cavitation both during and after superplastic forming have been developed and the effect of these treatments on the mechanical properties are discussed. 29 ref.—AA

Category: 31. Controlled Terms: Aluminum base alloys/Mechanical properties / Titanium base alloys/Mechanical properties/Superplastic forming/Strain rate/Tensile strength/Deformation effects/Fatigue (materials) / Deformation effects/ Crack propagation/Deformation effects/Cavitation / Microstructure. Alloys: 7475/8090/Supral 220/Supral 100/AL/Ti-6Al-4V / Ti-15V-3Cr-3Al-3Sn/TI

55-0234 METADEX Issue 8801
Diffusion Bonding in the Manufacture of Aircraft Structures.
Stephen, D.; Swadling, S. J. AGARD (NATO) July 1986; (CP-398); 7.1-7.17; ISSN 0365-2475; in English

Over the last twenty years, considerable Aerospace Reserach and Development effort has been directed to the development of the diffusion bonding (D.B.) process as a means of manufacture of low cost structures. To date the main thrust of these developments have been associated with titanium which has inherent metallurgical characteristics which make this material ideally suited for joining by this technique. For these titanium alloys which exhibit superplastic (SPF) properties, the combined processes of SPF and DB considerably extend the range of low cost and structural efficient titanium aerospace components which can be manufactured; even as replacements for conventionally fabricated aluminium alloy components. Recent developments in the SPF of high strength aluminiums and metal matrix composites has stimulated work in the field of DB of aluminium. It is thought that in the longer term this field of DB could have the highest levels of application. This paper details the range of aerospace structural forms which can and are currently being manufactured using the diffusion bonding process. The processes options, bond integrity, and NDT aspects are discussed. 18 ref.—AA

Category: 55. Controlled Terms: Titanium base alloys/Welding/Aluminum base alloys/Welding/Diffusion welding/ Superplastic forming/Aircraft components/Welding/Composite materials/Welding/Welded joints / Nondestructive testing / Quality control. Alloys: Ti-6Al-4V/TI

55-0231 METADEX Issue 8801
The Application of Diffusion Bonding and Laser Welding in the Fabrication of Aerospace Structures
Dunkerton, S. B.; Davies, C. J. AGARD (NATO) July 1986; (CP-398); 3.1-3.12; ISSN 0365-2475; in English

A review is given of work undertaken in both diffusion bonding and laser welding which is relevant to the aerospace industry. The wide use of superplastic forming/diffusion bonding of titanium alloys is mentioned with reference to particular applications. This is extended to include the newly developed superplastic aluminium alloys and data are presented on the diffusion bonding of conventional aluminium materials. The laser welding of aluminium, steel, nickel alloys and titanium alloys is covered with detail given on mechanical properties such as tensile and fatigue. The weld quality is shown to be tolerant to changes in process parameters by means of weldability lobes while dimensional tolerances such as beam/joint alignment and component fit-up can be critical. Finally, the development of laser beam spinning is mentioned with data on the increased tolerance to joint mismatch. 4 ref.—AA

Category: 55. Controlled Terms: Titanium base alloys/Welding/Aluminum base alloys/Welding/Nickel base alloys/ Welding/Diffusion welding/Laser beam welding/Development/Superplastic forming/Welded joints/Mechanical properties/Fatigue (materials)/Tensile strength/Weldability. Alloys: 5083 / 2014A/AL/Ti-6Al-4V/TI/C263/NI

8711-116856 Compendex
SUPERPLASTICITY FORMS A BOND WITH SHEET METAL PRODUCTS
Anon. Aust Mach Prod Eng v 40 n 2 Mar 1987 p 13-14 Coden: AMPDA ISSN: 0004-9719 In ENGLISH. Doc. Type: JOURNAL ARTICLE Treatment Des.: Applications EXPERIMENTAL

Two metallurgical phenomena that are being exploited increasingly in the forming of sheet metal and other components largely for the aerospace industries, yet remain relatively unused in more general applications, are superplasticity and diffusion bonding. Although both have found their way into processes in their own right, it is becoming more common to find both used in the production of parts more conventionally associated with pressing and spot welding or riveting. The article describes the combined process, and its application in the production of such things as one-piece tanks and membrane forming

Card-A-Lert Codes: 535 (Rolling, Forging and Forming)/531 (Metallurgy and Metallography)/Controlled Terms: (*SHEET AND STRIP METAL — *Forming) / (METALS AND ALLOYS — Superplasticity)/(BONDING — Diffusion)/ Uncontrolled Terms: SHEET METAL PRODUCTS

8709-95095 Compendex
SUPERPLASTICITY FORMS A BOND WITH SHEET METAL PRODUCTS
Anon. Eng Mater Des v 31 n 2 Feb 1987 p 33, 35 Coden: EMTDA ISSN: 0013-8045 In ENGLISH. Doc. Type: JOURNAL ARTICLE Treatment Des.: GENERAL REVIEW

Two metallurgical phenomena that are being exploited increasingly in the forming of sheet metal and other components largely for the aerospace industries, yet remain relatively unused in more general applications, are superplasticity and diffusion bonding. Although both have found their way into processes in their own right, it is becoming more common to find both used in the production of parts more conventionally associated with pressing and spot welding or riveting

Card-A-Lert Codes: 535 (Rolling, Forging and Forming)/531 (Metallurgy and Metallography)/Controlled Terms: (*SHEET AND STRIP METAL — *Forming) / (METALS AND ALLOYS — Superplasticity)/BONDING/Uncontrolled Terms: SHEET METAL PRODUCTS/SUPERPLASTIC FORMING/DIFFUSION BONDING

8708-85660 Compendex
MATERIAL CHARACTERIZATION OF SUPERPLASTICALLY FORMED TITANIUM (Ti-6Al-2Sn-4Zr-2Mo) SHEET
Ossa, William A.; Royster, Dick M. PRC Kentron Inc, Hampton, VA, USA. NASA Tech Pap 2674 May 1987 38p Coden: NTPAD ISSN: 0148-8341 In ENGLISH Refs: fs. Doc. Type: Report Review Treatment Des.: EXPERIMENTAL

The aerospace industry has focused considerable interest on the near-alpha titanium alloy Ti-6Al-2Sn-4Zr-2Mo (Ti6242) because of both its high-temperature properties and its superplastic forming (SPF) capabilities. This paper describes current research to characterize selected mechanical properties of Ti-6242 sheet in the SPF-strained condition, both with and without heat treatment, and compares the results with those obtained on as-received material. Tensile and creep tests were conducted, and metallographic analysis was performed to show the effect of 100 to 700 percent SPF strain on titanium properties. Analysis shows that as a result of SPF processing, both tensile and yield strengths, as well as elongation, are moderately reduced. Creep tests at 800 degree F and 1000 degree F show that the SPF processed material displays superior creep resistance compared with the as-received material. (Edited author abstract) 12

Card-A-Lert Codes: 542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/535 (Rolling, Forging and Forming)/421 (Strength of Materials; Mechanical Properties)/423 (Miscellaneous Properties and Tests of Materials) / Controlled Terms: (*TITANIUM AND ALLOYS — *High Temperature Effects)/(SHEET AND STRIP METAL — Mechanical Properties)/(METALS TESTING — Creep) / (STRESSES — Strain)/Uncontrolled Terms: TITANIUM SUPERPLASTIC FORMING/HIGH TEMPERATURE PROPERTIES

N87-29641/4/XAD NTIS
Superplasticity
Advisory Group for Aerospace Research and Development, Neuilly-sur-Seine (France).*National Aeronautics and Space Administration, Washington, DC (056102000 AD455458). Rept no: AGARD-LS-154, ISBN-92-835-1557-9, c1987, 201p, See also N87-29642. Lecture Series held at Wright-Patterson AFB, OH., September 24-25, 1987, In Luxembourg, September 28-29, 1987; and in London, England, September 28-29, 1987, NTIS Prices: PC A10/MF A01

No abstract available

Fld: 71N. Controlled terms: *Alloys/*Cavitation flow/*Conferences / *Diffusion welding/*Process control (Industry)/ *Superplasticity / Manufacturing/Metal sheets/Aluminum/Design analysis/Mechanical properties/Titanium. Uncontrolled terms: *Foreign technology/NTISNASAE / NTISFNFR

87A52700 NASA IAA Conference Paper Issue 24
Effect of Fe on superplastic Al-Li alloys
(AA)NAVROTSKI, G.; (AB)WARD, B. R. (AA)(Cornell University, Ithaca, NY); (AB)(Reynolds Metals Co., Metallurgy Laboratories, Richmond, VA). IN: Aluminum alloys: Their physical and mechanical properties; Proceedings of the International Conference, Charlottesville, VA, June 15-20, 1986. Volume 2 (A87-52651 24-26). Warley, England, Engineering Materials Advisory Services, Ltd., 1986, p. 1285-1299. Research supported by Reynolds Metals Co. 860000 p. 15 refs 20 In: EN (English) p.3853

In principle, sizeable reduction in weight of aerospace components can be accomplished by forming them superplastically from lighter weight, higher modulus aluminum-lithium alloys. An investigation into the effect of Fe and Cd on the superplastic properties of an Al-2.6Cu-2.4Li-0.2Zr alloy is reported. When tested at a constant true strain-rate, the Fe-and Cd-free alloys performed better than the Fe (0.07 percent)-or Cd (0.05 percent)-containing alloys. Under conditions of constant cross-head speed testing, however, the formability of the alloys reversed. Information on the flow stress, strain-rate sensitivity value, elongation to failure, true stress-strain relationships, microstructures, and variance of m is presented. Author

Category code: 26 (metallic materials). Controlled terms: *ALUMINUM ALLOYS /*IRON /*LITHIUM ALLOYS / *SUPERPLASTICITY/ CADMIUM/MICROSTRUCTURE/STRAIN RATE/STRESS-STRAIN RELATIONSHIPS /

87A49627 NASA IAA Meeting Paper Issue 22
Designing for low cost fabrication
Proceedings of the Workshop, Loughborough University of Technology, England, Apr. 16, 17, 1986 London, Royal Aeronautical Society, 1987, 233 p. For individual items see A87-49628 to A87-49637. 870000 p. 233 In: EN (English) $45.12 p.3527

The present conference discusses approaches to cost engineering, the improvement of cost effectiveness in fabrication of aerospace structures with conventional materials, the importance of research to aerospace parameters and concepts, airframe cost engineering methods, carbon fiber-reinforced composite applications to airframes, the application of composites to propulsion systems, and design criteria for low cost structures fabrication. Also discussed are low cost missile structure fabrication, the use of plastics for the fabrication of a launch-tube saddle, low cost composite rotor blade manufacture, novel manufacturing techniques for aircraft, kevlar manufacturing for weapons systems, novel airframe bonding techniques, and low cost aluminum and titanium alloy manufacture using superplastic forming and diffusion bonding. O.C.

Category code: 01 (aeronautics). Controlled terms: *AIRCRAFT DESIGN /*CONFERENCES /*COST EFFECTIVENESS /*DESIGN TO COST /*FABRICATION /*PRODUCTION COSTS/AIRCRAFT CONSTRUCTION MATERIALS/AIRCRAFT PRODUCTION COSTS/CARBON FIBERS/DESIGN ANALYSIS/FIBER REINFORCED COMPOSITES/LOW COST/WEAPON SYSTEMS/WEAPONS DEVELOPMENT /

87A48015 NASA IAA Conference Paper Issue 21
Superplastic forming of titanium — A production viewpoint
(AA)COMLEY, P. N. (AA)(Murdock, Inc., Compton, CA). IN: Titanium 1986; Products and applications; Proceedings of the International Conference, San Francisco, CA, Oct. 19-22, 1986. Volume 2 (A87-47976 21-26). Dayton, OH, Titanium Development Association, 1987, p. 1034-1059. 870000 p. 26. In: EN (English) p.3377

The current status of the superplastic forming (SPF) of titanium is reviewed with emphasis on applications in the aerospace industry. In particular, attention is given to the mechanism of SPF in titanium alloys, the advantages of SPF and the areas of application, considerations in SPF, and design criteria in SPF. The discussion also covers the tooling used in the SPF of titanium, process parameters, the forming equipment, and the manufacture of an SPF part. V.L.

Category code: 26 (metallic materials). Controlled terms: *AIRCRAFT CONSTRUCTION MATERIALS / *FORMING TECHNIQUES /*HOT WORKING /*METAL SHEETS /*SUPERPLASTICITY /*TITANIUM ALLOYS /*TOOLS/ALUMINUM ALLOYS/COST EFFECTIVENESS/GRAIN SIZE / THERMAL EXPANSION/VANADIUM ALLOYS /

87A48000 NASA IAA Conference Paper Issue 21
Titanium diffusion bonding in the manufacture of aircraft structure (AA)STEPHEN, D. (AA)(British Aerospace, PLC, Civil Aircraft Div., Bristol, England). IN: Titanium 1986; Products and applications; Proceedings of the International Conference, San Francisco, CA, Oct. 19-22, 1986. Volume 2 (A87-47976 21-26). Dayton, OH, Titanium Development Association, 1987, p. 603-630. 870000 p. 28 refs 21 In: EN (English) p.3376

The use of diffusion bonding (DB) and combinations of DB with other processes such as superplastic forming (SPF) in the manufacture of aerospace structures is reviewed, with emphasis on titanium alloy structures. In particular, attention is given to the theoretical and practical aspects of the basic DB process, the mechanical properties of DB joints, typical defects of DB joints, and nondestructive testing techniques. The discussion also covers the equipment used in DB processing, combined SPF/DB process, the integrity of DB and SPF/DB components, and the economic and performance advantages of DB and SPF/DB technology. V.L.

Category code: 26 (metallic materials) Controlled terms: *AIRCRAFT CONSTRUCTION MATERIALS / *DIFFUSION WELDING /* MECHANICAL PROPERTIES /*METAL BONDING /*SUPERPLASTICITY /*TITANIUM ALLOYS/FAILURE MODES/HIGH TEMPERATURE TESTS / METAL SHEETS/ NONDESTRUCTIVE TESTS/SHEAR STRENGTH/SURFACE ROUGHNESS /

8704-31890 Compendex
SUPERPLASTICITY IN AEROSPACE — ALUMINIUM
Pearce, Roger (Ed.); Kelly, Larry (Ed.). Cranfield Inst of Technology, Sch of Industrial Science, Cranfield, Engl. Superplast in Aerosp — Aluminium, Cranfield, Engl, Jul 12-15 1985 Publ 1985 463p In ENGLISH. Doc. Type: Conference Proceedings Treatment Des.: Applications THEORETICAL EXPERIMENTAL

This conference proceedings contains 23 papers. Various aspects of superplastic forming of aluminum and aerospace structures are covered. Some of the topics discussed are aircraft materials; airframes; aluminum sheet; alloy cavitation; metallography; and deformation mechanisms. Other topics covered include metal joining; welding; bonding; alloy composition effect; grain size effect; microstructures; and mechanical properties. Technical and professional papers from this conference are indexed and abstracted with the conference code no. 09130 in the Ei Engineering Meetings (TM) database produced by Engineering Information, Inc

Card-A-Lert Codes: 541 (Aluminum and Alloys)/535 (Rolling, Forging and Forming)/421 (Strength of Materials; Mechanical Properties)/415 (Metals, Plastics, Wood and Other Structural Materials)/652 (Aircraft)/538 (Welding and Bonding)/Controlled Terms: (*ALUMINUM AND ALLOYS — *Forming)/(METALS AND ALLOYS — Superplasticity)/

(AIRCRAFT MATERIALS — Aluminum)/(AIRCRAFT — Airframes)/(METAL FORMING — Cavitation)/(ALUMINUM SHEET — Forming) / Uncontrolled Terms: METAL JOINING/MECHANICAL PROPERTIES/METALLOGRAPHY / AEROSPACE STRUCTURES/EIREV

87A44746 NASA IAA Conference Paper Issue 19
Aluminum lithium alloys for aerospace
(AA)PEEL, C. J.; (AB)MCDARMAID, D. S. (AB)(Royal Aircraft Establishment, Materials and Structures Dept., Farnborough, England). IN: Materials in aerospace; Proceedings of the First International Conference, London, England, Apr. 2-4, 1986. Volume 2 (A87-44729 19-23). London, Royal Aeronautical Society, 1986, p. 348-372. 860000 p. 25 refs 10 In: EN (English) p.2973

The development status and prospects for Al-Li alloys applicable to aerospace structures are evaluated in light of data for various critical properties. Attention is given to the damage-tolerance, medium strength and high strength characteristics of the 2090, 2091, 8090, and 8091 Al-Li alloys, with emphasis on such problematic requirements as the achievement of high toughness in the damage-tolerant condition and of additional strength in sheets and forgings, in which the requisite nucleation of precipitation products is difficult. Results of recent corrosion and fatigue tests, as well as of studies of the properties of the 8090 alloy in sheet form after superplastic forming, are presented. O.C.

Category code: 26 (metallic materials). Controlled terms: *AEROSPACE INDUSTRY /*ALUMINUM ALLOYS / *LITHIUM ALLOYS /* MECHANICAL PROPERTIES /*WEIGHT REDUCTION / CORROSION PREVENTION/DAMAGE ASSESSMENT/FRACTURE STRENGTH/HIGH STRENGTH ALLOYS/STIFFNESS/ SUPERPLASTICITY /

87A44731 NASA IAA Conference Paper Issue 19
Materials in helicopters — A review
(AA)HOLT, D. (AA)(Westland, PLC, Yeovil, England). IN: Materials in aerospace; Proceedings of the First International Conference, London, England, Apr. 2-4, 1986. Volume 1 (A87-44729 19-23). London, Royal Aeronautical Society, 1986, p. 17-28. 860000 p. 12 In: EN (English) p.2935

The application of advanced materials to the construction of helicopter structures and mechanical components is motivated by the desire to improve airworthiness and safety, reduce empty weight fraction, reduce purchase and operating costs, and improve operational performance capability. In introducing a novel material, attention must be given to intrinsic variability in basic material properties and variability due to manufacturing processes, as well as to the behavior to be expected from the material in the course of use throughout its operational environment and service life. An examination is presently made of glass and carbon fiber-reinforced plastics, thermoplastic matrices, welded titanium alloy, gearbox case magnesium alloy castings, and superplastically formed structures. O.C.

Category code: 01 (aeronautics). Controlled terms: *AIRCRAFT CONSTRUCTION MATERIALS /*COMPOSITE MATERIALS /* HELICOPTER DESIGN/AEROSPACE INDUSTRY / ALUMINUM ALLOYS/ROTOR BLADES/ WEIGHT REDUCTION /

87A44729 NASA IAA Meeting Paper Issue 19
Materials in aerospace
Proceedings of the First International Conference, London, England, Apr. 2-4, 1986. Volumes 1 & 2. Conference sponsored by the Royal Aeronautical Society. London, Royal Aeronautical Society, 1986. Vol. 1, 243 p.; vol. 2, 213 p. For individual items see A87-44730 to A87-44750. 860000 p. 456 In: EN (English) Price of two volumes, $62 p.2962

The present conference on state-of-the-art applications of advanced materials in the aerospace industries considers topics in helicopter components, guided missiles, propulsion systems, spacecraft, and airframe primary structures. Attention is given to Ti alloy castings, thermoplastic matrix composites, microwave/IR transparencies, self-reinforcing polymers, carbon/ carbon composites, engine hot component ceramics, and mechanical alloying of Al-Mg-Li alloys. Also discussed are superalloy component durability enhancement methods, metal-matrix composites, the superplastic forming/diffusion bonding of Ti alloys, and Al-Li alloys for aerospace applications. O.C.

Category code: 23 (chemistry/materials). Controlled terms: *AIRCRAFT CONSTRUCTION MATERIALS / *COMPOSITE MATERIALS /* CONFERENCES /*HEAT RESISTANT ALLOYS/ALUMINUM ALLOYS/CARBON-CARBON COMPOSITES/CERAMICS/CORROSION PREVENTION/HELICOPTERS/ION IMPLANTATION/LITHIUM ALLOYS/LUBRICATION / MAGNESIUM ALLOYS/METAL SURFACES/ REINFORCED PLASTICS/SURFACE FINISHING / TITANIUM ALLOYS /

61-0601 METADEX Issue 8710
The British Aerospace Experimental Aircraft Programme and the Role of System Development Cockpits.
Whiteside, P. V. ICAS 1986: 15th Congress of the International Council of the Aeronautical Sciences. Vol. 1; London, UK; 7-12 Sept. 1986; American Institute of Aeronautics and Astronautics; 1633 Broadway, New York, New York 10019, USA; 1986; 8710-72-0531; 201-212; in English

An overview of the BAe experimental aircraft programme (EAP), including a resume of previous work that has been contributory to its success, is presented. Emphasis is placed on the systems aspects of the aircraft, outlining the architecture and materials employed including carbon fibre composite wing skins and spars, superplastic formed/diffusion bonded Ti structures used in between the engines and Al—Li alloy used in the flaperon skins, the design and software production philosophy, finally focusing on the system development and test philosophy. Included in the final section is a description of the EAP development cockpit and other complimentary facilities that played a major role in the design of the EAP cockpit.—AA

Category: 61. Controlled Terms: Aluminum base alloys/Titanium/Aircraft components/Materials selection/Military planes/Materials selection

51-1619 METADEX Issue 8710
Superplastic Forming — a New Possibility for Light Alloys.
Green, D. Lamiera Apr. 1986; 23, (4); 112-113; ISSN 0391-5891; in Italian

Based on the recent practice of the British Aerospace Company principles, applications and advantages of superplastic forming are presented. Titanium alloys (type Ti—6Al—4V) present an excellent formability, obtaining both weight-and production cost-reduction. Superplastic forming is combined with diffusion bonding for efficient use in the aerospace industry. Superplastic forming of Al and Al alloys is briefly outlined, claiming good results for Supral type alloy (Superform Metals). Possibilities of using Al—Li alloys are pointed out.—K.Z.

Category: 51. Controlled Terms: Titanium base alloys/Metal working/Aluminum base alloys/Metal working/ Superplastic forming/Diffusion welding / Economics/Aerospace. Alloys: Ti-6Al-4V/TI

54-0745 METADEX Issue 8708
Development of Particulate Reinforced High Strength Aluminum Alloy for Aerospace Applications.
Nair, C. G. K.; Krishnadev, M. R.; Dutta, D. Advanced Materials Technology '87; Anaheim, California, USA; 6-9 Apr. 1987; Society for the Advancement of Material and Process Engineering; P.O. Box 2459, Covina, California 91722, USA; 1987; 8708-72-0443; 889-901; in English

A SiC particulate reinforced high strength Zr refined Al—6Zn—2.3Mg—1.8Cu alloy has been developed through the powder metallurgy route. Powder prepared by gas atomization is mixed with SiC particulates (approx = 10 mu m) canned in vacuum, followed by hot pressing and extrusion. Microstructure and properties of the composite are discussed. The resulting alloy is forgeable and heat-treatable to develop high strength. The matrix can be made superplastic by thermomechanical treatment and this makes the composite more versatile. The alloy is economical to produce and is considerably cheaper than Al—Li alloys. The development, processing, properties and potential applications of the alloy are discussed. Both optical and scanning electron microscope have been used to study the microstructure and fracture. 4 ref.—AA

Category: 54. Controlled Terms: Aluminum base alloys/Powder technology / Silicon carbide/Alloying additive/ Dispersion hardening alloys/Alloy development/Powder metallurgy parts/Alloy development/Hot pressing / Thermomechanical treatment / Aircraft components/Materials selection/Gears / Materials selection/Automotive engines/ Materials selection. Alloys: Al-6Zn-2.3Mg-1.8Cu/AL

31-3447 METADEX Issue 8708
Material Characterization of Superplastically Formed Titanium (Ti—6Al—2Sn—4Zr—2Mo) Sheet. (Pamphlet)
Ossa, W. A.; Royster, D. M. NASA Tech. Pap. NASA TP-2674; May 1987; Pp 35; in English

The aerospace industry has focused considerable interest on the near-alpha Ti alloy, Ti—6Al—2Sn—4Zr—2Mo (Ti-6242) because of both its high-temperature properties and its superplastic forming (SPF) capabilities. Current research to characterize selected mechanical properties of Ti-6242 sheet in the SPF-strained condition, both with and without heat treatment is outlined, and compared with the results of those obtained on as-received material. Tensile and creep tests were conducted, and metallographic analyses performed to show the effect of 100-700% SPF strain on Ti properties. Analysis shows that as a result of SPF processing, both tensile and yield strengths, as well as elongation, are moderately reduced. Creep tests at 800 deg C and 1000 deg F (427 and 538 deg C) show that the SPF processed material displays superior creep resistance compared with the as-received material. A post-SPF duplex-anneal heat treatment had no beneficial effect on tensile and creep properties.

Category: 31. Controlled Terms: Titanium base alloys/Mechanical properties / Sheet metal/Mechanical properties/ Tensile properties/Deformation effects / Creep (materials)/Deformation effects/Superplastic forming. Alloys: Ti-6Al-4Zr-2Sn-2Mo/Ti-6242/TI

87A25709 NASA IAA Journal Article Issue 10 **On the transition from superplastic to non-superplastic deformation at high strain rates**
(AA)CHOKSHI, ATUL H.; (AB)MUKHERJEE, AMIYA K. (AB)(California, University, Davis) AF-AFOSR-86-00575
Scripta Metallurgica (ISSN 0036-9748), vol. 20, Dec. 1986, p. 1771-1774. 861200 p. 4 refs 18 In: EN (English) p.1403

The role of subgrain size in the transition at high strain rates from the superplastic region II to the nonsuperplastic region III was critically examined, using a modified equation of Bird et al. (1969), which relates the subgrain size to the applied stress. The theoretical predictions were compared with literature data on the following materials: Zn-0.1 pct Ni-0.05 pct Mg, Cu-2.8 pct Al-1.8 pct Si-0.4 pct Co, Mg-34 pct Al, and Pb-62 pct Sn. The results suggest that the transition at high strain rates is not caused solely by the development of subgrains with dimensions equal to the grain size. Deformation at high strain rates in region III is attributed to intergranular dislocation creep, with significant contributions from grain boundary sliding. It is suggested that superplasticity can be extended to higher strain rates, by decreasing the grain size and by adding suitable elements that will influence the appropriate diffusivities in regions II and III. I.S.

Category code: 26 (metallic materials). Controlled terms: *EUTECTIC ALLOYS /*GRAIN BOUNDARIES /*GRAIN SIZE /*PLASTIC DEFORMATION /*STRAIN RATE /*SUPERPLASTICITY/AEROSPACE INDUSTRY/ ELONGATION/MECHANICAL PROPERTIES / MICROSTRUCTURE /

TIB/B86-80862/XAD NTIS
Konstruieren mit superplastisch umgeformten Blechbauteilen in der Luft-und Raumfahrt (Designing with Super-Elastic Formed Sheet Metal Parts in Aerospace and Space Travel)
Winkler, P.J.; Maerkis, F. Messerschmitt-Boelkow-Blohm G.m.b.H., Munich (Germany, F.R.). Information und Dokumentation (064776008) Rept no: MBB-BB—581/84-Oe, 1984, 17p, In German. VDI conference on designing with sheet metal, Essen (Germany, F.R.), 8-9 Nov 1984, Also published in VDI-Berichte no. 523 (1984) p. 207-223. NTIS Prices: PC E07

Some high strength aluminium or titanium alloys important for aerospace and space travel have superelastic properties. Superelastic forming, whose technical and economic advantages are described, makes it possible to manufacture complicated formed components. This expands the design possibilities. Suitable metal forming and manufacture of parts is described with the example of some helicopter parts. (RHM). (Copyright (c) 1986 by FIZ. Citation no. 86:080862.)

Fld: 71N, 51C, 84G, 84C. Controlled terms: *Mechanical properties/*Titanium alloys/*Aluminium alloys/ Superplasticity/Elongation/Plastic deformation / Eutectic alloys/Aircraft design/Metal sheets. Uncontrolled terms: *Foreign technology/NTISFIZ/NTISFNGE/NTISLNGER

N87-17060/1/XAD NTIS
Application du Soudage Par Diffusion Associe au Formage Superplastique (SPF/DB) a la Realisation de Structures en Toles Minces de Ta6v (Application of Diffusion Bonding Combined with Superplastic Forming (SPF/DB) to the Fabrication of Tagv Metal Sheet Structures)
Boire, M.; Jolys, P. Societe Nationale Industrielle Aerospatiale, Suresnes (France). Lab. Central.*National Aeronautics and Space Administration, Washington, DC (071739001 SQ456835). Jul 86, 14p, Text in French. In AGARD Advanced Joining of Aerospace Metallic Materials 14p, NTIS Prices: (Order as N87-17051 PC A12/MF A01)

The potential applications of combined superplastic forming and diffusion bonding (SPF/DB) technology to the fabrication of TA6V titanium alloy structures are discussed. The conditions for the application of this technology are summarized as well as the impacts of certain material properties. Examples of structural fabrication processes are presented and an economical technique is suggested. Finally, the main problems which need to be resolved for a fully-developed industrial process are examined

Fld: 41F, 94G. Controlled terms: *Diffusion welding/*Forming techniques / *Metal sheets/*Metal-metal bonding/ *Superplasticity/*Titanium alloys / Economic factors/Fabrication/Plastic deformation/Pressure dependence / Production engineering / Temperature dependence. Uncontrolled terms: *Foreign technology/NTISNASAE/NTISLNFRE/NTISFNFR

52-0941 METADEX Issue 8706
Application of Aluminum Alloy Superplasticity in Aerospace.
Matsuo, M. J. Jpn. Inst. Light Met. Jan. 1986; 36, (1); 43-50; ISSN 0451-5994; in Japanese

A review of the application of Al alloy superplasticity in aerospace was made. Superplastic forming (SPF) could save 50% weight, reduce the size of parts and decrease the maintenance costs. The typical superplastic Al alloys, Supral 100, 220 and Al—Li alloys (8090) are described. The maximum elongation for the alloys was 500-700%. The cavitation and the influences of grain size and intermetallic compounds during superplastic forging are discussed. Cavitation could be eliminated by high temperature hydraulic pressure after SPF. The diffusion bonding of the SPF Al alloys is described. The parts of the fighter planes and helicopters which could be processed with SPF are indicated. 23 ref.—X.S.

Category: 52. Controlled Terms: Aircraft components/Metal working/Aluminum base alloys/Metal working/ Superplastic forming/Cavitation/Hot isostatic pressing/Fatigue life. Alloys: Supral 220/8090/AL

51-1098 METADEX Issue 8706
Progress, Problems and the Future in Rapid Solidification Technology.
Grant, N. J. The Second International Conference on Iron and Steel Technology and New Materials; Pohang, South Korea; 7-9 Oct. 1986; Pohang Iron and Steel Co. (POSCO); Pohang, South Korea; 1986; 8706-72-0360; 173-198; in English

Because of perceived, urgent, near term applications for Al alloys and for Ni-base superalloys, it is not suprising to find that funding for these two classes of alloys was more abundant and developments in alloys, processes and properties more spectacular than for ordinary steels, copper base alloys, etc. These are alloys which have an important bearing on aircraft and aerospace applications. In the course of the development of rapid solidification technology (RST) and processing (RSP), developments which are still extremely young, one would expect that parallel developments would take place in alloy design and alloy processing; in techniques for achieving the desired high quenching and high solidification rates; and in learning to live with a spectrum of new problems such as highly increased reactivity due to fine powder sizes, new types and sources of contamination, new consolidation methods, etc. In spite of the emphasis on Al and superalloy materials, there has been sufficient work done with high speed tool steels, stainless steels, Cu base alloys, magnesium alloys and low alloy steels to be convinced that RST benefits are not reserved to the high alloy materials, the high priced materials, the unusual or peculiar compositions, etc. Excellent results have been achieved in almost every class of alloy, clearly with greater or lesser degrees of benefit depending on the properties of interest. 28 ref.—AA

Category: 51. Controlled Terms: Aluminum base alloys/Casting/Superalloys / Casting/Steels/Casting/Copper base alloys/Casting/Magnesium base alloys/Casting/Rapid solidification/Mechanical properties/Cooling effects/Superplasticity. Alloys: 2024/Al-4.5Cu-1.6Mg-1Li / Al-4.5Cu-1.6Mg-3Li/Al-4.5Cu-1.6Mg-1.6Li/X2020/7075/Al-6.7Mg-1.6Li/AL / IN100/Mar M509/NI/SP/316/Fe-0.05C-17Cr-12Ni-2.5Mo-0.3Ti / Fe-0.15C-17Cr-12Ni-2.5Mo-0.9Ti/Fe-0.25C-17Cr-12Ni-2.5Mo-1.5Ti/SSA / Fe-0.03C-23.5Cr-5Ni-1Cu-1.5Mo/Fe-0.04C-23Cr-7Ni-1Cu-1.5Mo / Fe-0.09C-27Cr-5.5Ni-1Cu-1.5Mo-0.11Nb/Fe-25Cr-20Ni-12B/SSD

87N20959# NASA STAR Conference Proceedings Issue 14
Net shape technology in aerospace structures. Volume 3. Appendix. Emerging Net Shape Technologies.
Presentations of a workshop held on March 27-29, 1985 in Santa Barbara, California/Final Report, 1984 — 1986
(AA)STEINBERG, MORRIS A. National Academy of Sciences — National Research Council, Washington, D. C. (NB210513) Committee on Net Shape Technology in Aerospace Structures. AD-A176510 F49620-85-C-0107 861200 p. 526 Workshop held in Santa Barbara, Calif., 27-29 Mar. 1985 In: EN (English) Avail: NTIS HC A23/MF A01 p.1843

This report is in four volumes. Papers presented by invited speakers at the workshops appear in Volume 3 (emerging technologies). This document is an appendix to Net Shape Technology in Aerospace Structures, Vol. 1. It contains 30 reports by representatives of industry on emerging net shape technologies for the fabrication of aerospace parts. Technologies include: powder metallurgy, coatings, superplastic forming/diffusion bonding, hot isostatic pressing, ceramic-ceramic composites. These reports were presented at a workshop held March 27 to 29, 1985 in Santa Barbara, California. GRA

Category code: 01 (aeronautics). Controlled terms: *AEROSPACE VEHICLES /*AIRCRAFT STRUCTURES / *ALUMINUM ALLOYS /*CERAMICS /*COATINGS /*CONFERENCES /*DIFFUSION WELDING /*FABRICATION /*HEAT RESISTANT ALLOYS /*HOT PRESSING /*ISOSTATIC PRESSURE /*TITANIUM ALLOYS/AEROSPACE SYSTEMS/CALIFORNIA / COMPOSITE MATERIALS/COMPOSITE STRUCTURES/ POWDER METALLURGY/PRODUCTION ENGINEERING/SHAPES /

87N20957# NASA STAR Technical Report Issue 14
Net shape technology in aerospace structures. Volume 1/Final Report, 1984 — 1986
(AA)STEINBERG, MORRIS A. National Academy of Sciences — National Research Council, Washington, D. C. (NB210513) Committee on Net Shape Technology in Aerospace Structures. AD-A176508 F49620-85-C-0107 861100 p. 117 In: EN (English) Avail: NTIS HC A06/MF A01 p.1843

This report is in four volumes. Volume 1 is the committee's assessment of the state of net shape technology for aerospace applications based on briefings and discussion at the workshops. This report is an assessment of current and possible applications of net shape technologies by the Air Force, including: precision forging of alloys, powder metallurgy, structural ceramics, superplastic forming, diffusion bonding, vapor deposited coatings, etc.; and composites, including organic matrix composites, metal matrix composites, ceramic and carbon matrix composites in the manufacture of aircraft components. It also includes road maps of research and development efforts in performance and manufacturing technologies and resource allocation. GRA

Category code: 01 (aeronautics). Controlled terms: *AEROSPACE VEHICLES /*AIRCRAFT EQUIPMENT / *AIRCRAFT STRUCTURES /*ALUMINUM ALLOYS /*COMPOSITE MATERIALS /*COMPOSITE STRUCTURES /* CONFERENCES /*MANUFACTURING /*MATRIX MATERIALS /*METAL MATRIX COMPOSITES /* POWDER METALLURGY /*PRODUCTION ENGINEERING /*RESOURCE ALLOCATION /*SHAPES/ AEROSPACE SYSTEMS/ARMED FORCES (UNITED STATES)/CARBON/CERAMICS/COATINGS/DIFFUSION WELDING/FORGING / HEAT RESISTANT ALLOYS/NETS/NICKEL ALLOYS/ORGANIC MATERIALS/STEELS / TITANIUM ALLOYS/VAPOR DEPOSITION /

31-2670 METADEX Issue 8706
Superplasticity in an Al—Li—Cu—Mg—Zr Alloy.
Otsuka, M.; Tsurumaki, K.; Niimura, M.; Horiuchi, R. J. Jpn. Inst. Light Met. Nov. 1986; 36, (11); 752-758; ISSN 0451-5994; in Japanese

Superplastic behavior of an Al—Li—Cu—Mg—Zr alloy (Lital A) for aerospace application has been investigated over the temperature range 600-800K and the initial strain rate range 1 x 10 exp —4 -1 x 10 exp —2 s exp —1 . In cold-rolled and annealed sheet specimens having fine and equiaxed grains of approx 5 mu m diameter, superplasticity was observed › 725K, wherein the maximum elongation of 480% appeared and the strain rate sensitivity was determined to be 0.45 at 800K. As-rolled sheet specimens became much more superplastic than annealed ones, when holding time prior to the tensile test was chosen appropriately. No improvement in ductility was found in the temperature increasing tensile tests. Extruded specimens having fine but elongated grain structure did not show superplasticity at all, probably because grain boundary sliding was suppressed due to a geometrical reason. 22 ref.—AA

Category: 31. Controlled Terms: Aluminum base alloys/Mechanical properties / Superplasticity/Lithium/Alloying elements. Alloys: Lital A / Al-2.7Li-1.2Cu-0.9Mg-0.14Zr/Al-2.4Li-1.2Cu-07Mg-0.14Zr/AL

57-0471 METADEX Issue 8705
Application of Stop-Off Coating by Ion Plating.
Turner, B. J. British Aerospace 15 Oct. 1986; 26 Mar. 1986; in English. Patent no: GB2173511A; UK

A method is disclosed of applying a dense tightly adhering stop-off coating to the surface of a substrate including steps of introducing the substrate into an inert atmosphere maintained at a low pressure and depositing on the surface the stop-off material by means of an ion-plating process. The substrate is a metal blank part of an article to be produced by a plastic forming and diffusion bonding process or it may be a mould tool for use in a superplastic forming process. The stop-off material is preferably yttria boron nitride, graphite or alumina, and the coating has a thickness of the order of 2 mu m. The stop-off coating prevents bonding of adjacent metal parts during forming or diffusion bonding.

Category: 57. Controlled Terms: Ion plating/Inert atmospheres/Protective coatings/Diffusion welding/Forming

54-0465 METADEX Issue 8705
Titanium NNS Technology Shaping Up.
Kubel, E. J. Jr. Adv. Mater. Process. inc. Met. Prog. Feb. 1987; 131, (2); 46-50; ISSN 0026-0665; in English

Developments regarding near-net-shape (NNS) processing of Ti alloys for aerospace structures are discussed. The processes include precision casting, powder metallurgy (PM) fabrication, precision forging and superplastic forming/diffusion bonding (SPF/DB). Specific materials mentioned include Ti—6Al—4V sandwich structures, Ti—6Al—6V—2Sn, Ti—10V—2Fe—3Al, Ti—5Al—2.5Sn and Ti—8Al.—J.M.S.

Category: 54. Controlled Terms: Titanium base alloys/Powder technology / Powder metallurgy parts/Sintering (powder metallurgy)/Cold pressing/Hot isostatic pressing/Precision forging/Superplastic forming

87N17056 NASA STAR Conference Paper Issue 09
Diffusion bonding in the manufacture of aircraft structure
(AA)STEPHEN, D.; (AB)SWADLING, S. J. British Aerospace Aircraft Group, Bristol (England). (BX274903) Civil Aircraft Div. In AGARD Advanced Joining of Aerospace Metallic Materials 17 p (SEE N87-17051 09-37) 860700 p. 17 In: EN (English) Avail: NTIS HC A12/MF A01 p.1183

Over the last twenty years, considerable aerospace research and development effort has been directed to the development of the diffusion bonding (DB) process as a means of manufacture of low cost structures. To date the main thrust of these developments have been associated with titanium which has inherent metallurgical characteristics which make this material ideally suited for joining by this technique. For these titanium alloys which exhibit superplastic properties, the combined processes of superplastic forming (SPF) and DB considerably extend the range of low cost and structurally efficient titanium aerospace components which can be manufactured; even as replacements for conventionally fabricated aluminum alloy components. Recent developments in the SPF of high strength aluminums and metal matrix composites has stimulated work in the field of DB of aluminum. It is thought that in the longer term this field of DB could have the highest levels of application. This paper details the range of aerospace structural forms which can and are currently being manufactured using the diffusion bonding process. The process options, bond integrity, and nondestructive test (NDT) aspects are discussed. Author

Category code: 37 (mechanical engineering). Controlled terms: *AIRCRAFT PRODUCTION /*AIRCRAFT STRUCTURES /*ALUMINUM ALLOYS /*DIFFUSION WELDING /*MANUFACTURING /*METAL-METAL BONDING /*TITANIUM ALLOYS /* WELD STRENGTH / DEFECTS/ECONOMIC FACTORS/FORMING TECHNIQUES/NONDESTRUCTIVE TESTS / SUPERPLASTICITY /

87N17054# NASA STAR Conference Paper Issue 09
The application of diffusion bonding and laser welding in the fabrication of aerospace structures
(AA)DUNKERTON, S. B.; (AB)DAWES, C. J. Welding Inst., Cambridge (England). (WM313175) Solid Phase Welding Group. In AGARD Advanced Joining of Aerospace Metallic Materials 12 p (SEE N87-17051 09-37) 860700 p. 12 In: EN (English) Avail: NTIS HC A12/MF A01 p.1182

A review is given of work undertaken in both diffusion bonding and laser welding which is relevant to the aerospace industry. The wide use of superplastic forming/diffusion bonding of titanium alloys is mentioned with reference to particular applications. This is extended to include the newly developed superplastic aluminum alloys and data are presented on the diffusion bonding of conventional aluminum materials. The laser welding of aluminum, steel, nickel alloys and titanium alloys is covered with detail given on mechanical properties such as tensile and fatigue. The weld quality is shown to be tolerant to changes in process parameters by means of weldability lobes while dimensional tolerances such as beam/joint alignment and component fit-up can be critical. Finally, the development of laser beam spinning is mentioned with data on the increased tolerance to joint mismatch. Author

Category code: 37 (mechanical engineering). Controlled terms: *AEROSPACE ENGINEERING /*ALUMINUM ALLOYS /*DIFFUSION WELDING /*FABRICATION /*LASER WELDING /*TITANIUM ALLOYS /*WELD STRENGTH /*WELDED JOINTS / METAL FATIGUE / MISALIGNMENT/NICKEL ALLOYS/STEELS/ SUPERPLASTICITY/TENSILE STRENGTH /

87N17051# NASA STAR Conference Proceedings Issue 09
Advanced Joining of Aerospace Metallic Materials (conference proceedings)
Advisory Group for Aerospace Research and Development, Neuilly-Sur-Seine (France). (AD455458) Structures and Materials Panel. AGARD-CP-398; ISBN-92-835-0397-X; AD-A173979 Loughton, England 860700 p. 270 Meeting held in Oberammergau, West Germany, 11-13 Sep. 1985 In ENGLISH and FRENCH In: AA (Mixed) Avail: NTIS HC A12/MF A01 p.1182

The papers contained in this report provide a review of the state-of-the-art of advanced joining techniques currently available to the manufacturers of aerospace equipment and identify newly emerging techniques and joining problems. Computational weld mechanics, diffusion bonding and superplastic forming, electron-beam welding, tungsten inert gas welding, inspection methods, and repair techniques are addressed. For individual titles see N87-17052 through N87-17071.

Category code: 37 (mechanical engineering) Controlled terms: *AEROSPACE ENGINEERING /*CONFERENCES / *DIFFUSION WELDING /* ELECTRON BEAM WELDING /*GAS TUNGSTEN ARC WELDING /*METAL-METAL BONDING /*WELD TESTS /*WELDED JOINTS / AIRCRAFT MAINTENANCE/AIRCRAFT PRODUCTION/ENGINE PARTS/HEAT RESISTANT ALLOYS/INSPECTION/NONDESTRUCTIVE TESTS/ ULTRASONIC FLAW DETECTION /

87N17032# NASA STAR Conference Paper Issue 09
Diffusion welding of component parts in the aviation and space industries
(AA)DISAM, J.; (AB)MIETRACH, D. Royal Aircraft Establishment, Farnborough (England). (R2785060) RAE-TRANS-2147 Transl. into ENGLISH of conference paper "Diffusionsschweissen von Bauteilen in der Luft-und Raumfahrt" presented at the International Conference on Brazing, High-Temperature Brazing and Diffusion Welding, 1981 860400 p. 12 Conference held in Duesseldorf, West Germany, 21-22 Sep. 1981; sponsored by German Welding Society In: EN (English) Avail: NTIS HC A02/MF A01 p.1179

Investigations were performed and successfully completed on the materials combinations titanium/titanium, steel/steel, titanium/aluminum, titanium/permen- orm, and aluminum/aluminum. Experience gained during the ZKP flap-track program, as well as during the manufacturing of heat pipes for space technology, was used for diffusion bonding of further components such as titanium/steel connections for TV-satellites. Superplastically formed/diffusion bonded titanium components, as well as different material combinations will be used to an increasing extent for civil and military aircraft in the future. In this context, examples are given from civil and military programs. Comments are made on materials and materials combinations. Costs are compared with costs for components manufactured by conventional methods. Trends in the USA show that in the future increasing emphasis will be laid on the joining of semi-finished products having quasi-final contours such as titanium castings and powder-metallurgical parts by, for example, diffusion bonding in order to cut the structural costs in airframe construction drastically. Author

Category code: 37 (mechanical engineering). Controlled terms: *AEROSPACE ENGINEERING /*AIRCRAFT STRUCTURES /*COST EFFECTIVENESS /*DIFFUSION WELDING /*MECHANICAL PROPERTIES /*POWDER METALLURGY /* STRUCTURAL DESIGN/AIRFRAMES / ALLOYS/CIVIL AVIATION/HEAT PIPES /

61-0228 METADEX Issue 8704
Aluminium—Lithium Alloys for Aerospace.
Peel, C. J.; McDarmaid, D. S. Materials in Aerospace. The First International Conference. Vol. II; London, UK; 2-4 Apr. 1986; The Royal Aeronautical Society; 4 Hamilton Place, London W1V 0BQ, UK; 1986; 8704-72-0229; 348-372; in English

The current situation in the development of a series of lightweight Al—Li alloys is reviewed by examining the more critical properties of the most developed of the emerging alloys. Appropriate data for alloys 2090, 2091, 8090 and 8091 are considered in three main categories, damage tolerant, medium strength and high strength. Particular emphasis is placed on problem areas, such as the achievement of high toughness in the damage tolerant conditions and on obtaining extra strength in sheet and forgings, where nucleation of precipitation is difficult. Recent corrosion and fatigue results are reviewed together with some studies of the properties of sheet 8090 after superplastic forming. 10 ref.—AA

Category: 61. Controlled Terms: Aeronautical engineering/Aluminum base alloys / Alloy development/Lithium/ Alloying elements/Damage/Tensile properties/Corrosion resistance/Superplastic forming. Alloys: 2090/2091/8090/ 8091/AL.

61-0209 METADEX Issue 8704
Superplastic Forming and Diffusion Bonding of Titanium.
Stephen, D. Designing With Titanium; Bristol, UK; 7-9 July 1986; The Institute of Metals; 1 Carlton House Terrace, London SW1Y 5DB, UK; 1986; 8704-72-0223; 108-124; in English

The processes of superplastic forming (SPF) and diffusion bonding (DB) are becoming widely accepted within the airframe manufacturing business as processes which have a potential to save weight and cost in the manufacture of a range of airframe structures. Because of its inherent superplastic and diffusion bonding properties, coupled with its attractive strength and corrosion properties, the major activities in this field have been focussed on the use of the Ti—6Al—4V alloy. British Aerospace has a considerable background in the development of these processes and has now a wide experience in the design and manufacture of production components. This note draws upon this experience in discussing the advantages that can result. 6 ref.—AA

Category: 61. Controlled Terms: Airframes/Fabrication/Titanium base alloys / Fabrication / Superplastic forming/ Diffusion welding. Alloys: Ti-6Al-4V/TI

87N10994 NASA STAR Conference Paper Issue 02
International Conference — Superplasticity in Aerospace Aluminium
(AA)BROAD, K. S. (AA)comp. British Aerospace Aircraft Group, Bristol (England). (BX274903) Advanced Manufacturing Research Dept. BAE-S85/AMR/0066; ETN-86-97945 850723 p. 12 Conference held in Bedford, England, 12-15 Jul. 1985 In: EN (English) Avail: Issuing Activity p.184

The use of superplastic aluminum in aerospace was discussed. Superplastic forming to produce cost effective lightweight parts was considered. Limitations from microstructural fractures and resultant properties were covered, particularly the problem of cavitation. Special processing techniques and the evolution of alloys to overcome limitations were treated. ESA

Category code: 26 (metallic materials). Controlled terms: *AEROSPACE ENGINEERING /*AIRCRAFT CONSTRUCTION MATERIALS /* ALUMINUM /*SPACECRAFT CONSTRUCTION MATERIALS / *SUPERPLASTICITY/CAVITATION FLOW/COST EFFECTIVENESS / MICROSTRUCTURE/PRODUCTION ENGINEERING/TECHNOLOGY UTILIZATION/WEIGHT REDUCTION /

62-0008 METADEX Issue 8701
Materials for Aerospace.
Steinberg, M. A. Sci. Am. Oct. 1986; 255, (4); 66-72; ISSN 0036-8733; in English

New developments in materials and materials processing are giving aerospace component designers a wider range of selection. Advanced materials currently in use include a variety of metal-and plastic-matrix composites (Al matrix, Ti matrix, graphite/epoxy, graphite/polyimide, C/C), improved Al alloys and ceramics, rigid-rod polymers, and high-strength steels. Promising new alloy and component fabrication methods include rapid-solidification technology, superplastic forming, diffusion bonding, and net shape fabrication. The Office of Science and Technology (OST) recently submitted a program outlining desirable material characteristics for the aerospace industry that are expected to be achieved by the year 2000. The program calls for increased fuel efficiency through weight reduction made possible by expected developments in ceramics, Al—Li alloys, and advanced composites.—G.P.K.

Category: 62. Controlled Terms: Aluminum base alloys/Composite materials / Titanium base alloys/Composite materials/Fiber composites/Mechanical properties/Aircraft components/Materials selection/Aerospace engines / Materials selection/Strength to weight ratio

46-0015 METADEX Issue 8701
Advanced Metals.
Kear, B. H. Sci. Am. Oct. 1986; 225, (4); 158-167; ISSN 0036-8733; in English

Recent developments in advanced metals, with emphasis on aerospace applications, are reviewed. Metals considered include Ni-and cobalt-based superalloys, titanium aluminide intermetallics, and Ti-based alloys. Advanced fabrication procedures include directional solidification, superplastic forming, mechanical alloying, and rapid solidification. The micromechanisms responsible for the ability of advanced metals to resist deformation better than conventional metals are described.—G.P.K.

Category: 46. Controlled Terms: Nickel base alloys/Alloy development/Cobalt base alloys/Alloy development/ Intermetallics/Alloy development/Titanium base alloys/Alloy development/Aerospace engines/Materials selection / Aircraft components/Materials selection/Turbine blades/Materials selection

31-0442 METADEX Issue 8701
Superplasticity in Aluminium Alloys.
Baba, Y.; Yoshida, H. J. Jpn. Soc. Technol. Plast. Mar. 1986; 27, (302); 333-338; ISSN 0038-1586; in Japanese

The Lloyd conditions for superplasticity in aluminium alloys are discussed. Properties of Al—Zn eutectic alloys, AlZnCa co-crystallisation alloy, AlCu and AlMg (e.g. Supral 100), AlZnMgCu (7075, 7475), AlLi and aluminium SiC composites are compared and discussed. Uses of superplastic alloys are exemplified by 7475 as in single part formation for aerospace use. Problems discussed are cavity formation, hot processing strain and welding. 31 ref.—M.O.

Category: 31. Controlled Terms: Aluminum base alloys/Mechanical properties / Superplasticity/Superplastic forming/ Aircraft components/Weldability. Alloys: Al-78Zn/ZN/7475/8090/Supral 220/2004/Neopral/AL / Al-5Zn-5Ca/Supral 100/7075/AL

31-0303 METADEX Issue 8701

Effect of Iron on Superplastic Al—Li Alloys.

Navrotski, G.; Ward, B. R. Aluminum Alloys: Their Physical and Mechanical Properties. Vol. II; Charlottesville, Virginia, USA; 15-20 June 1986; Engineering Materials Advisory Services Ltd.; 339 Halesowen Rd., Cradley Heath, Warley, W. Midlands B64 6PH, UK ; 1986; 8701-72-0043; 1285-1299; in English

In principle, sizeable reductions in weight of aerospace components can be accomplished by superplastically forming them from lighter weight, higher modulus Al—Li alloys. An investigation into the effect of Fe and Cd on the superplastic properties of an Al—2.6Cu—2.4Li—0.2Zr alloy is reported. When tested at a constant true strain-rate, the Fe and Cd free alloys performed better than the Fe (0.07%) or Cd (0.05%) containing alloys. Under conditions of constant crosshead speed testing, however, the formability of the alloys reversed. Information on the flow stress, strain-rate sensitivity value, elongation to failure, true stress—strain relationships, microstructures and variance of m is presented. 20 ref.—AA

Category: 31. Controlled Terms: Aluminum base alloys/Mechanical properties / Superplasticity/Alloying effects/Iron/ Alloying additive/Cadmium / Alloying additive/Stress strain curves/Lithium/Alloying elements / Aircraft components/Materials selection. Alloys: Al-2.6Cu-2.4Li-0.2Zr / Al-2.6Cu-2.42Li-0.19Zr-0.02Fe/Al-2.62Cu-2.42Li-0.20Zr-0.07Fe / Al-2.47Cu-2.41Li-0.20Zr-0.01Fe-0.05Cd/AL / Al-2.7Cu-2.46Li-0.18Zr-0.07Fe-0.05Cd/AL

8611-104917 Compendex

ADVANCED TECHNOLOGY ALUMINUM MATERIALS FOR AEROSPACE APPLICATIONS

Pritchett, T. R. Kaiser Aluminum & Chemical Corp, Pleasanton, CA, USA. Light Met Age v 44 n 7-8 Aug 1986 p 10-14 Coden: LMAGA ISSN: 0024-3345 In ENGLISH Refs: 100 refs. Doc. Type: JOURNAL ARTICLE Treatment Des.: Applications HISTORICAL

New materials developed by the aluminum industry to combat high fuel costs and the inroads of composites include rapidly solidified (atomized, melt spun, strip or roller quenched) alloys, light weight Al-Li alloys and composites (metal matrix, resin/polymer laminates). Limited improvements in performance of conventional alloys continue to be made by impurity control, minor element additions, and better heat treatment and thermomechanical working practices. New manufacturing techniques such as superplastic forming of complex shapes, precision die forging and one-piece non-critical casting are also being used to reduce the weight and cost of some components

Card-A-Lert Codes: 541 (Aluminum and Alloys)/652 (Aircraft)/655 (Spacecraft) / 415 (Metals, Plastics, Wood and Other Structural Materials)/542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/549 (Nonferrous Metals and Alloys in General)/Controlled Terms: (*ALUMINUM AND ALLOYS — *Aerospace Applications)/(AEROSPACE VEHICLES — Materials)/(AIRCRAFT MATERIALS — Light Metals)/(ALUMINUM LITHIUM ALLOYS — Aerospace Applications) / Uncontrolled Terms: RAPID SOLIDIFCATION

8611-116165 Compendex

SUPERPLASTIC FORMING AND DIFFUSION BONDING OF TITANIUM ALLOYS

Ghosh, A. K.; Hamilton, C. H. Rockwell Int Science Cent, Thousand Oaks, CA, USA. Def Sci J v 36 n 2 Apr 1986 p 153-177 Coden: DSJOA ISSN: 0011-748X In ENGLISH Refs: fs. Doc. Type: JOURNAL ARTICLE Treatment Des.: Applications GENERAL REVIEW

Among the high strength materials of interest in the aerospace industry, titanium holds a position of foremost importance. It is a powerful element for construction of space vehicles, and commercial and military aircrafts. Since corrosion has proved to be an expensive factor in aircraft maintenance, titanium is a highly desirable material in the fabrication of key aircraft structural components. New and advanced fabrication methods for titanium components are emerging today to replace age-old fabrication processes and reduce component cost. Superplastic forming and diffusion bonding are two such advanced fabrication technologies which when applied individually or in combination can provide significant cost and weight benefits and a rather broad manufacturing technology base. This paper briefly reviews the state of understanding of the science and technology of superplastic forming of titanium alloys, and their diffusion bonding capability. Emphasis has been placed on the metallurgy of superplastic flow in two phase titanium alloys, the microstructural and external factors which influence this behavior. (Edited author abstract) 36

Card-A-Lert Codes: 542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/535 (Rolling, Forging and Forming)/404 (Civil Defense and Military Engineering)/531 (Metallurgy and Metallography)/652 (Aircraft) / 655 (Spacecraft)/Controlled Terms: (*TITANIUM AND ALLOYS — *Forming) / (TITANIUM METALLOGRAPHY — Diffusion)/TITANIUM METALLURGY/(AIRCRAFT — MILITARY — Materials)/(AEROSPACE VEHICLES — Materials)/BONDING / Uncontrolled Terms: DIFFUSION BONDING/ELEVATED TEMPERATURE DUCTILITY / SUPERPLASTIC FORMING

86A46826 NASA IAA Meeting Paper Issue 22

Rapidly solidified powder aluminum alloys

Proceedings of the Symposium, Philadelphia, PA, April 4, 5, 1984 (AA)FINE, M. E.; (AB)STARKE, E. A., JR. (AA)ED.; (AB)ED. (AA)(Northwestern University, Evanston, IL); (AB)(Virginia, University, Charlottesville). Symposium sponsored by ASTM. Philadelphia, PA, American Society for Testing and Materials (ASTM Special Technical Publication, No. 890), 1986, 552 p. For individual items see A86-46827 to A86-46849. 860000 p. 552 In: EN (English) Members, $51.20; nonmembers, $64 p.0

Topics discussed include the need for rapidly solidified PM Al alloy development for aerospace and land vehicle applications, the solidification theory, and stereological characterization of porous microstructures. Papers treating the characterization of rapidly solidified materials, rapid solidification of highly undercooled Al powders, Al3Li precipitate modification in an Al-Li-Zr alloy, hardening mechanism in rapidly solidified Al-8Fe alloy, the fabrication of high-strength PM extrusions, hypereutectic Al-based alloys, the properties of Al alloys with 8-12 wt pct Fe and Al-Li-Be alloys, the microstructure of supercooled submicrometer Al-Cu alloy powder, and the influence of hydrogen on the ductility of 7091 and 7090 PM alloys. Attention is given to PM processing of Al alloy 7091, dynamic compaction of rapidly solidified Al alloy powders, the effects of alloy chemistry on superplastic forming of rapid solidification-processed Al-Li alloys, thermomechanical treatment of 2124 PM Al alloys, and a method for degassing evaluation of Al PM alloys. Studies concerned with high strength PM alloys, a technique for assessing the corrosion properties of Al PM alloys, the development of dispersion-strengthened Al alloys, the microstructure/mechanical property relationships for thermal treatments of Al-Cu-Mg-X PM Al alloys, PM and IM Al alloy forgings, the annealing behavior and tensile properties of elevated-temperature PM Al alloys, the effect of compositional changes on the microstructure and properties of rapidly solidified Al3-Li alloys and the use of rapid solidification for the development of 2XXX alloys. I.F.

Category code: 26 (metallic materials). Controlled terms: *ALUMINUM ALLOYS /*CONFERENCES /*METAL POWDER /*POWDER METALLURGY /*RAPID QUENCHING (METALLURGY) / CORROSION RESISTANCE/ HEAT TREATMENT/HIGH STRENGTH ALLOYS/INTERMETALLICS / LOW DENSITY MATERIALS/MICROSTRUCTURE/POROUS MATERIALS/SCALING LAWS / TERNARY ALLOYS /

86A40251 NASA IAA Meeting Paper Issue 19
Aluminium-lithium alloys III; Proceedings of the Third International Aluminium-Lithium Conference, Oxford University, England, July 8-11, 1985
(AA)BAKER, C.; (AB)GREGSON, P. J.; (AC)HARRIS, S. J.; (AD)PEEL, C. J. (AA)ED.; (AB)ED.; (AC)ED.; (AD)ED. (AA)(Alcan International, Ltd., Banbury, England); (AB)(Southampton, University, England); (AC)(Nottingham University, England); (AD)(Royal Aircraft Establishment, Farnborough, England)

Conference organized and sponsored by the Institute of Metals. London, Institute of Metals, 1986, 640 p. For individual items see A86-40252 to A86-40315. 860000 p. 640 In: EN (English) Members, $63.20; nonmembers, $79 p.2760

The papers presented in this volume focus on the fundamental metallurgy, production and processing aspects, and aerospace applications of Al-Li alloys. Emphasis is placed on mechanical properties and fabrication parameters in comparison with the existing aerospace alloys as well as requirements and specific applications of the alloys in airframes and engines in fixed-wing aircraft and helicopters. Papers are included on the production of aluminum-lithium alloys with high specific properties, fatigue crack propagation in mechanically alloyed Al-Li-Mg alloys, superplastic aluminum-lithium alloys, fundamental aspects of hardening in Al-Li and Al-Li-Cu alloys, and mechanical properties of Al-Li-Zn-Mg alloys. V.L.

Category code: 26 (metallic materials). Controlled terms: *ALUMINUM ALLOYS /*CONFERENCES /*LITHIUM ALLOYS /*MECHANICAL PROPERTIES /*METALLURGY/CRACK PROPAGATION/HEAT TREATMENT/ MAGNESIUM ALLOYS / METAL FATIGUE / MICROSTRUCTURE/POWDER METALLURGY/PRECIPITATION HARDENING/PRODUCT DEVELOPMENT/RAPID QUENCHING (METALLURGY) /

86A31475 NASA IAA Meeting Paper Issue 13
Materials and processes; Proceedings of the Fifth Technology Conference, Montreux, Switzerland, June 12-14, 1984. Volumes 1&2
Conference sponsored by the Society for the Advancement of Material and Process Engineering. Geneva, Switzerland, Society for the Advancement of Material and Process Engineering, 1985. Vol. 1, 185 p.; vol. 2, 191 p. No individual items are abstracted in these volumes. 850000 p. 376 In: EN (English) Price of two volumes, $51 p.1812

The present conference on advanced aerospace materials gives attention to high performance thermoplastic matrix composites and their manufacturing techniques, the filament winding of complex components, the mechanized manufacture of FRP components for aircraft secondary structures, a novel high strain-to-failure prepreg, the manufacture of accurate glass and carbon fiber preforms for resin injection, a helicopter composite tail unit, Al-Li alloys, the durability of Arall, the potential weight savings obtainable in future transport aircraft through the use of advanced materials, superplastically formed Ti and Al alloys for aerospace applications, and assembly bonding with room temperature-curing adhesives. Also discussed are novel composite systems for use in primary aircraft structures, accelerated moisture absorption in carbon-epoxy, the thermoanalytic characterization of matrix resins and composites, the definition of microstructures in hybrid reinforced plastics, ceramic components for automotive powerplants, Kevlar-reinforced automotive components, metallized textile fabrics and their applications, and the role of S-2 glass fibers in advanced composites. O.C.

Category code: 23 (chemistry/materials). Controlled terms: *COMPOSITE MATERIALS /*CONFERENCES / *MATERIALS SCIENCE/AIRCRAFT CONSTRUCTION MATERIALS/ALUMINUM ALLOYS/AUTOMOBILE ENGINES/CARBON FIBER REINFORCED PLASTICS/FIBER REINFORCED COMPOSITES/FILAMENT WINDING/GLASS FIBER REINFORCED PLASTICS / GRAPHITE-EPOXY COMPOSITES/HELICOPTER TAIL ROTORS/KEVLAR (TRADEMARK) MATRIX MATERIALS/METAL MATRIX COMPOSITES/METALLIZING/ MOISTURE CONTENT/POLYMER MATRIX COMPOSITES/SUPERPLASTICITY/TEXTILES/ THERMOPLASTIC RESINS/TITANIUM ALLOYS /

86A31309 NASA IAA Journal Article Issue 13
Superplastic Al-Cu-Li-Mg-Zr alloys
(AA)WADSWORTH, J.; (AB)LEWIS, R. E.; (AC)PELTON, A. R. (AB)(Lockheed Research Laboratories, Palo Alto, CA); (AC)(DOE, Ames Laboratory, IA). F33615-78-C-5203 Metallurgical Transactions A — Physical Metallurgy and Materials Science (ISSN 0360-2133), vol. 16A, Dec. 1985, p. 2319-2332. Research supported by the Lockheed Independent Research and Development Program. 851200 p. 14 refs 52 In: EN (English) p.1826

The Al-Cu-Li-Mg-Zr alloys in which superplastic properties have been developed have been manufactured by both ingot and powder metallurgy methods. After conventional manufacturing, the alloys are not superplastic but rather require further thermomechanical processing. Attention is given to the microstructural changes that occur during this processing as functions of temperature, strain rate, and processing history. Room temperature properties in the solution-treated and peak-aged condition are noted for all the alloys and compared with those of non-Li admixture superplastic alloys, giving emphasis to properties of interest in aerospace applications, such as specific modulus, specific strength, and a buckling failure criterion. O.C.

Category code: 26 (metallic materials). Controlled terms: *ALUMINUM ALLOYS /*COPPER ALLOYS /*HIGH STRENGTH ALLOYS /* LITHIUM ALLOYS /*POWDER METALLURGY /*SUPERPLASTICITY/CHEMICAL COMPOSITION/ELECTRON MICROSCOPY/INGOTS / MAGNESIUM ALLOYS/SOLIDIFICATION/ SPACECRAFT STRUCTURES/THERMOMECHANICAL TREATMENT/ZIRCONIUM ALLOYS /

61-0527 METADEX Issue 8609
Materials and Manufacturing in Aerospace.
Clementson, A. ME—Proceedings of the Second Conference on Materials Engineering; London, UK; 5-7 Nov. 1985; Mechanical Engineering Publications; P.O. Box 24, Northgate Avenue, Bury St Edmunds, Suffolk IP32 6BW, UK; 1985; 8609-72-0367; 189-194; in English

In 1979 British Aerospace embarked on major development initiatives in new materials for airframes. An important aim was to achieve a close collaboration between production and design interests. Three key areas were selected; carbon fibre, titanium superplastic forming and diffusion bonding, and advanced aluminium alloys. The programmes are outlined and contrasted with reference to the process of introducing new materials and manufacturing processes.—AA

Category: 61. Controlled Terms: Airframes/Alloy development/Aluminum base alloys/Carbon fibers/Materials selection/Titanium base alloys

55-1902 METADEX Issue 8609
Quick Take-Off for SPFDB Titanium Fabrication.
Weld. Met. Fabr. Mar. 1985; 53, (2); 48-49; ISSN 0043-2245; in English

After three years, superplastic forming and diffusion bonding (SPFDB) techniques are being used successfully by the British Aerospace process. The process is described and includes silk screening a bond inhibitor, diffusion bonding at 940 deg C, and inflating with inert gas for three-dimensional parts. Expansion in civil, military, and missile industries is described for the process and several parts are pictured. Cost savings are described for many parts.—J.C.J.

Category: 55. Controlled Terms: Aircraft components/Blow molding/Diffusion welding/Superplastic forming/Titanium base alloys/Welding

31-3368 METADEX Issue 8609
Structural Materials in Aeronautics: Prospects and Perspectives. II.
Pope, G. G. Aerospace (London) May-June 1986; 13, (5); 22-30; ISSN 0001-933X; in English

AL—Li alloys for aeronautics applications are being studied in three grades—medium strength, high strength and fatigue resistant. A composition has been found which meets the medium strength requirements and a plant built to produce slabs and billets up to 3000 kg. Advantages of Li-containing alloys over conventional metallic materials and CFRP are indicated. Rapid solidification methods for Al alloys are being investigated. Al—Cr—Fe alloys with 5-9% Cr and 0.6-3% Fe have been produced by evaporation—condensation and have considerably improved tensile properties over precipitation hardening systems. Al—Cr—Fe is very thermally stable and corrosion resistant. Potential of 7000 series Al alloys for superplastic forming is being studied. Materials for gas turbine blades are discussed. Consideration is given to unidirectionally solidified and single crystal materials; advanced ceramics such as Si—Al—0—N and carbon fibres in C matrix; coatings to provide oxidation-and thermal resistance; lightweight high-strength materials; and Ti-and Ni-based disc alloys. 24 ref.—E.J.S.

Category: 31. Controlled Terms: Aerospace engines/Airframes/Aluminum base alloys/Fatigue (materials)/Fiber composites/Materials substitution / Mechanical properties/Stiffness/Tensile strength/Titanium base alloys / Weight reduction. Alloys: XXXA/XXXB/7010/2024/CM001-1C/AL/Ti-6Al-4V / IMI550/IMI679/IMI685/IMI829/IMIEX 834/TI

52-1239 METADEX Issue 8608
Aluminum—Titanium Alloy. Lighter Construction Materials by Superplastic Forming.
Ind.-Anz. 28 Jan. 1986; 108, (8); 19; ISSN 0019-9036; in German

The use of Ti—6Al—4V for aircraft and aerospace structural parts can reduce weight by 7-10%. To use Ti as a substitute for Al, superplastic forming and diffusion welding are used in fabrication. As applied by British Aerospace, a 6 m high hydraulic press is used which can apply a force of 5000 MN to a 1.8 x 1.2 m press table. Cold shapes are preheated in a furnace, removed to the press and handled by robots, and hot-formed. Consideration of the use of Al alloys (Al—Li) in automobile manufacture indicates their cost to be an undeniable disadvantage relative to steel and plastics.—B.L.

Category: 52. Controlled Terms: Aircraft components/Aluminum base alloys / Metal working/Press forming/ Superplastic forming/Titanium base alloys

31-3135 METADEX Issue 8608
Developments in Sheet Metal.
Pearce, R. Sheet Met. Ind. Apr. 1986; 63, (4); 188, 190, 192; ISSN 0037-3435; in English

New materials and processes in sheet metals are reviewed. Strengthening of steel while maintaining formability can be achieved by solid-solution, grain size refinement, precipitation, dislocation structure and transformation product methods. Some properties of new steels are tabulated. A new superplastic Al—Li—Cu—Mg alloy has been developed for the aerospace industry, with 10% weight saving and increase in elastic modulus, which does not cavitate and can be diffusion bonded. The large increase in ductiity and elongation resulting from superplastic forming is explained. The method is used by Superform Metals Ltd. for Al—Cu—Zr (2004) and Al—Cu—Zr—Mg—Si (Supral 100, 220 and 5000), and in USA for Al—Cu—Mg—Ni—Fe alloys, e.g. 7475. Shape memory behaviour in which martensitic transformation is reversible over a small temperature range äoccurs in the Cu—Zn system, the temperature varying as Si, Al, Ga, Sn etc. are added. Applications include hydraulic control line couplers on USN fighters. 5 ref.—M.G.deM.

Category: 31. Controlled Terms: Aluminum base alloys/Carbon steels / Formability/High strength steels/Mechanical properties/Metal working / Shape memory alloys/Sheet metal/Superplastic forming/Titanium base alloys Alloys: 2004/ Supral 100/7475/Supral 220/Supral 5000/AL/Ti-6Al-4V / TI/CR1/CR2/SCL/350HS/400HS/SAHS/Ni-50Ti/NI/TI

31-3028 METADEX Issue 8608
Superplastic Aluminum—Lithium Alloys.
Henshall, C. A.; Nieh, T. G.; Wadsworth, J. Aluminium—Lithium Alloys III; Oxford, UK; 8-11 July 1985; The Institute of Metals; 1 Carlton House Terrace, London SW1Y 5DB, UK; 1986; 8608-72-0312; 199-212; in English

Superplasticity is now an accepted manufacturing technique for Ti-, Ni-and Al-based alloys. Recent work at Lockheed Palo Alto Research Laboratory has led to the development of superplasticity in a number of novel Al—Li-based alloys manufactured by ingot metallurgy and powder (rapid solidification) metallurgy techniques. These alloys are of considerable current interest in the aerospace and aircraft industries because lithium improves the modulus and decreases the density of aluminum. Considerations of design of Al—Li alloys for superplastic properties are discussed and the interactions of compositional requirements with thermomechanical processing are described. The superplastic properties of a range of Al—Li alloys are presented and a detailed comparison with Al-7475 alloy and commercial Al—Cu—Zr based alloys is made. The ultrafine microstructures of the Al—Li alloys are illustrated and the interrelationship of these microstructures with the high strain rates at which superplasticity is found on the alloys is highlighted. Phenomenological equations for superplastic flow have been developed for several of the alloys. Finally, room temp. properties of the alloys are presented and a comparison is made with other superplastic Al-base alloys. 15 ref.—AA

Category: 31. Controlled Terms: Alloying elements/Aluminum base alloys / Lithium/Materials substitution/Mechanical properties/Superplasticity. Alloys: Al-3Li-0.5Zr/Al-4Cu-3Li-0.5Zr/Al-3Cu-2Li-1Mg-0.2Zr / Al-2.5Li-1.5Cu-1Mg-0.1Zr/ AL/Al-2.5Cu-2.2Li-0.1Zr/Al-2.5 Li-1.1Cu-0.7Mg-0.14Zr/Al-2.2Cu-2.1Li-0.8Mg-0.16Zr/Al-4Cu-2Li-0.15Zr-0.04Cd / Al-2.6Cu-2.4Li-0.2Zr/AL

61-0394 METADEX Issue 8607
Forming of Stiffened Panels.
Beezley-Long, P. W.; Irwin, D. J.; Mansbridge, M. H.; Norton, J. British Aerospace, 21 Nov. 1985; 3 May 1985; in English. Patent no: EP0161892; EUR

A method is disclosed of forming a stiffened panel from first and second metal sheets, at least the first sheet being capable of both superplastic deformation and diffusion bonding, and also provided with at least one control region of different thickness compared with other regions of the sheet. The method includes the steps of: (i) attaching the sheets together at a series of attachment lines across their faces, the attachment lines and the control region or regions being in predetermined relationship with one another, placing the attached sheets in a mould and heating to within that temperature range within which superplastic deformation and diffusion bonding takes place; (ii) urging those areas of the first sheet between the attachment lines away from the second sheet by a common differential pressure at a rate within that range of strain rates at which superplastic deformation occurs to form a series of cavities between the two sheets such that peripheral parts of those areas urged away from the second sheet form side walls of neighbouring cavities and become diffusion bonded together to provide internal stiffeners of the finished panel; (iii) the control region or regions effecting local modification of the rate of superplastic deformation such that the internal stiffeners adopt a desired configuration and location.

Category: 61. Controlled Terms: Diffusion welding/Fabrication/Holes/Panels / Stiffening/Strain rate/Superplastic forming

86A22128 NASA IAA Meeting Paper Issue 08
Evolution of Aircraft
Aerospace Structures and Materials Symposium, Dayton, OH, April 24, 25, 1985, Proceedings.Symposium sponsored by AIAA. Dayton, OH, American Institute of Aeronautics and Astronautics, 1985, 137 p. For individual items see A86-22129 to A86-22144. 850000 p. 137 In: EN (English) $25 p.0

Various papers on the evolution of aircraft and aerospace structures and materials are presented. The topics addressed include: XB-70 structures and materials advances, structural evolution from B-58 to F-16, advanced composites in construction of the Beech Starship, structural and material considerations for advanced fighters, the evolution of reciprocating engines at Lycoming, aircraft design from the myth of make-do to Mach 3, and the Wright Brothers' experience in the evolution of aircraft design, structures and materials. Also considered are: evolution of the turbofan aircraft engine, X-15 high-temperature advanced structure, X-20 structures overview, ASSET program for technology development, Shuttle Orbiter airframe, airframe design to achieve minimum cost, superplastically formed-diffusion bonded titanium technology transition case study, transition of advanced materials and structures in single crystal blades, and composites technology transfer and transition. C.D.

Category code: 01 (aeronautics). Controlled terms: *AIRCRAFT CONSTRUCTION MATERIALS /*AIRCRAFT STRUCTURES /* CONFERENCES /*SPACECRAFT CONSTRUCTION MATERIALS /*SPACECRAFT STRUCTURES/AEROSPACE TECHNOLOGY TRANSFER/AIRCRAFT DESIGN/AIRCRAFT ENGINES/ AIRFRAMES/COMMERCIAL AIRCRAFT/COMPOSITE MATERIALS/COMPOSITE STRUCTURES/DIFFUSION WELDING/FIGHTER AIRCRAFT/SPACE SHUTTLE ORBITERS/SUPERPLASTICITY/WINGS/X-15 AIRCRAFT/X-29 AIRCRAFT /

8602-8972 Compendex
Flexible and Cost-Efficient Stamping of Small Batches of Body Parts FLEXIBLE UND KOSTENGUENSTIGE FERTIGUNG VON KLEINSERIEN IM KAROSSERIEBAU
Richards, John D.; Sawle, Rodger. Automobiltech Z v 87 n 7-8 Jul-Aug 1985 p 359-361 Coden: AUTZA ISSN: 0001-2785 In GERMAN. Doc. Type: JOURNAL ARTICLE

Stampings made from aluminum sheet using the superplastic method offer advantages whenever complex shapes have to be manufactured in small quantities. Using the superplastic aluminum alloys available today, nearly all requirements regarding the mechanical characteristics and stability of body parts can be fulfilled. The advantages of this manufacturing method lie in low tooling costs involved and the fact that the dies used are subjected to only slight mechanical stress. These advantages are helping make superplastic stampings better known in areas outside aerospace, where they had already proven their worth over the years. One of these newer areas is automobile manufacturing, where interesting potential is opening up for use of this method in prototypes, small batches, customized versions or similar applications. This report from Superform Metals of Worcester Ltd., England, shows that extensive experience has been gained in manufacturing parts for aircraft and automobile bodies. (Author abstract) In German

Card-A-Lert Codes: 535 (Rolling, Forging and Forming)/541 (Aluminum and Alloys)/415 (Metals, Plastics, Wood and Other Structural Materials)/662 (Automobiles and Smaller Vehicles)/652 (Aircraft)/Controlled Terms: (*ALUMINUM SHEET — *Forming)/(METAL FORMING — Stamping)/AUTOMOBILE MATERIALS/AIRCRAFT MATERIALS/ Uncontrolled Terms: SUPERPLASTIC STAMPINGS / SUPERPLASTIC METHOD

86A20037# NASA IAA Journal Article Issue 07
Research on high-strength aerospace aluminum alloys
(AA)STALEY, J. T. (AA)(Aluminum Company of America, Alloy Technology Div., Pittsburgh, PA). (CASI, Annual General Meeting, 31st, Ottawa, Canada, May 28, 1984) Canadian Aeronautics and Space Journal (ISSN 0008-2821), vol. 31, March 1985, p. 14-29. 850300 p. 16 refs 19 In: EN (English) p.1024

The utilization of aluminum alloys in airframe designs is discussed. The present aim in aircraft design is to save weight by decreasing density and increasing strength while maintaining corrosion resistance and increasing toughness, modulus, and fatigue resistance. Aluminum-lithium alloys and wrought powder metallurgy alloys are currently being studied for aircraft structures. Wrought powder metallurgy alloys have 30 percent higher strength than conventional alloys with improved toughness and corrosion. The application of dispersion hardened alloys for temperatures in the range 350-600 F and wrought powder metallurgy alloys for aircraft wheels is described. The fabrication and fatigue testing of the wrought powder metallurgy alloys are analyzed. The development of cost effective, lightweight, and resistant aircraft structures from metal matrix composites, aramid aluminum laminates, and superplastic 7475 is being investigated. I.F.

Category code: 26 (metallic materials). Controlled terms: *AIRCRAFT DESIGN /*AIRFRAME MATERIALS / *ALUMINUM ALLOYS /* HIGH STRENGTH ALLOYS/COST EFFECTIVENESS/FRACTURE STRENGTH/HEAT RESISTANT ALLOYS/LITHIUM ALLOYS / METAL MATRIX COMPOSITES/POWDER METALLURGY /

86A14474# NASA IAA Preprint Issue 03
Metals and plastics — State of the art and perspectives
Metaux et plastiques — Etat actuel et perspectives
(AA)BRANDT, J.; (AB)KELLERER, H.; (AC)WINKLER, P. (AC)(Messerschmitt-Boelkow-Blohm GmbH, Ottobrunn, West Germany). MBB-Z-49-85-OE Association Aeronautique et Astronautique de France, Journee des Pionniers Europeens, Paris, France, Apr. 25, 1985, Paper. 35 p. In French. 850400 p. 35 In: FR (French) p.280

Progress and expected advances in metals and plastics for aerospace usage, especially aircraft primary structures, are discussed. Metals, mostly Al and Ti alloys, are still the preferred materials, and are expected to be enhanced to have lower densities and display greater fracture resistance in the future. The advances will be achieved with new additives to the alloys, powder metallurgy, and superplastic forming. Metal matrix composites are also under investigation, as are fiber-reinforced plastics (FRPs), the latter being more developed and cheaper to produce than the metal counterparts. Kevlar and carbon fiber reinforcements are the most promising reinforcements; however, matrix materials have yet to reach reliability levels which would permit flight-certification of FRPs for primary structures. M.S.K.

Category code: 27 (nonmetallic materials). Controlled terms: *AIRCRAFT CONSTRUCTION MATERIALS / *ALUMINUM ALLOYS /*METAL MATRIX COMPOSITES /*REINFORCED PLASTICS /*TECHNOLOGY ASSESSMENT /*TITANIUM ALLOYS/AEROSPACE ENGINEERING / CARBON FIBER REINFORCED PLASTICS/FIBER REINFORCED COMPOSITES/KEVLAR (TRADEMARK)/POWDER METALLURGY / SUPERPLASTICITY / TECHNOLOGICAL FORECASTING /

86N11247*# NASA STAR Conference Paper Issue 02
Advances in joining techniques used in development of SPF/DB titanium sandwich reinforced with metal matrices
(AA)FISCHLER, J. E. Douglas Aircraft Co., Inc., Long Beach, Calif. (D1957175). In NASA. Langley Research Center Welding, Bonding and Fastening, 1984 p 297-322 (SEE N86-11227 02-23) 850900 p. 26 In: EN (English) Avail.: NTIS HC A21/MF A01 p.202

Three and four-sheet expanded titanium sandwich sheets have been developed at Douglas Aircraft Company, a division of McDonnell Douglas Corporation, under contract to NASA Langley Research Center. In these contracts, spot welding and roll seam welding are used to join the core sheets. These core sheets are expanded to the face sheets and diffusion bonded to form various type cells. The advantages of various cell shapes and the design parameters for optimizing the wing and fuselage concepts are discussed versus the complexity of the spot weld pattern. In addition, metal matrix composites of fibers in an aluminum matrix encapsulated in a titanium sheath are aluminum brazed successfully to the titanium sandwich face sheets. The strength and crack growth rate of the superplastic-formed/diffusion bonded (SPF/DB) titanium sandwich with and without the metal matrix composites are described. Author

Category code: 37 (mechanical engineering). Controlled terms: *ALUMINUM /*COMPOSITE STRUCTURES / *CRACK PROPAGATION /* DIFFUSION WELDING /*FIBER COMPOSITES /*HONEYCOMB CORES /*METAL MATRIX COMPOSITES /*SANDWICH STRUCTURES /*SHEATHS /*SPOT WELDS /*STRUCTURAL MEMBERS /* SUPERPLASTICITY /*THIN PLATES /*TITANIUM ALLOYS /*WING PANELS/AEROSPACE INDUSTRY/ MECHANICAL PROPERTIES / SPACECRAFT CONSTRUCTION MATERIALS /

86N11245*# NASA STAR Conference Paper Issue 02
Evaluation of superplastic forming and weld-brazing for fabrication of titanium compression panels
(AA)ROYSTER, D. M.; (AB)BALES, T. T.; (AC)DAVIS, R. C. National Aeronautics and Space Administration. Langley Research Center, Hampton, Va. (ND210491). In its Welding, Bonding and Fastening, 1984 p 253-270 (SEE N86-11227 02-23) 850900 p. 16 refs 0 In: EN (English) Avail.: NTIS HC A21/MF A01 p.202

The two titanium processing procedures, superplastic forming and weld brazing, are successfully combined to fabricate titanium skin stiffened structural panels. Stiffeners with complex shapes are superplastically formed using simple tooling. These stiffeners are formed to the desired configuration and required no additional sizing or shaping following removal from the mold. The weld brazing process by which the stiffeners are attached to the skins utilize spot welds to maintain alignment and no additional tooling is required for brazing. The superplastic formed/weld brazed panels having complex shaped stiffeners develop up to 60 percent higher buckling strengths than panels with conventional shaped stiffeners. The superplastic forming/ weld brazing process is successfully scaled up to fabricate full size panels having multiple stiffeners. The superplastic forming/weld brazing process is also successfully refined to show its potential for fabricating multiple stiffener compression panels employing unique stiffener configurations for improved structural efficiency. Author

Category code: 37 (mechanical engineering). Controlled terms: *AEROSPACE INDUSTRY /*BRAZING / *COMPRESSION LOADS /* DIFFUSION WELDING /*FABRICATION /*FLOW MEASUREMENT /*REINFORCEMENT (STRUCTURES) /*SKIN (STRUCTURAL MEMBER) /*STIFFENING /*SUPERPLASTICITY /*TENSILE STRENGTH /* TITANIUM ALLOYS /*WEBS (SUPPORTS) / MECHANICAL PROPERTIES/PLASTIC FLOW/SPACECRAFT CONSTRUCTION MATERIALS /

8601-265 Compendex
SUPERPLASTIC Al-Cu-Li-Mg-Zr ALLOYS
Wadsworth, J.; Pelton, A. R.; Lewis, R. E. Lockheed Palo Alto Research Lab, Palo Alto, CA, USA. Metall Trans A v 16A n 12 Dec 1985 p 2319-2332 Coden: MTTAB ISSN: 0360-2133 In ENGLISH Refs: fs. Doc. Type: JOURNAL ARTICLE *Treatment Des.:* EXPERIMENTAL

Superplastic properties have been developed in Al-Cu-Li-Mg-Zr alloys. The alloys have low Zr levels (<=0. 2 wt pct) and are experimental compositions originally designed as low-density, high-modulus, and high strength alloys for room temperature aerospace structural applications. The alloys have been manufactured both by an ingot metallurgy and a powder metallurgy (PM) route involving rapid solidification. After manufacturing, the alloys are not superplastic but require further thermomechanical processing. The microstructural changes that occur are described. Superplastic properties have been evaluated as a function of temperature, strain rate, and processing history. (Edited author abstract) 52

Card-A-Lert Codes: 541 (Aluminum and Alloys)/544 (Copper and Alloys)/542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/549 (Nonferrous Metals and Alloys in General)/536 (Powder Metallurgy)/421 (Strength of Materials; Mechanical Properties)/Controlled Terms: (*ALUMINUM COPPER LITHIUM MAGNESIUM ALLOYS — *Superplasticity)/(POWDER METALLURGY — Aluminum)/(AIRCRAFT MATERIALS — Light Metals)/Uncontrolled Terms: RAPID SOLIDIFICATION/THERMOMECHANICAL TREATMENT/HIGH TEMPERATURE PROPERTIES / MECHANICAL PROPERTIES/STRAIN RATE SENSITIVITY

71-0045 METADEX Issue 8602

Applications of Titanium and Titanium Alloys.

Farthing, T. W. Titanium—Science and Technology. Vol. 1; Munich, FRG; 10-14 Sept. 1984; Deutsche Gesellschaft fur Metallkunde; Adenauerallee 21, D-6370 Oberursel 1, FRG ; 1985; 8602-72-0069; 39-54; in English

Improvements in high temp. creep resistant alloys and development of fabrication techniques such as superplastic forming and diffusion bonding have led to increased use of Ti in th aerospace industry. Examples of this discussed include the Rolls Royce wide chord fan blade made from IMI 318, engine fan discs and blades from 6Al—4V, alpha—beta alloys for low temp. operation, Ti-6242 high pressure compressor discs, high temp. creep resistant alloys operating up to 600 deg C, Ti aluminides in the U.S., use of pure Ti in airframes, complex high stress structures such as the Panavia Tornado wing box, fine-grained alpha—beta alloys such as Ti—6Al—4V for complex components, and attempts to improve powder metallurgy parts. Consideration is given to general industrial applications, including heat exchangers, surgical implants, electrochemistry, and consumer products.—E.J.S.

Category: 71. Controlled Terms: Aerospace engines/End uses/Powder metallurgy parts/Titanium/Titanium base alloys. Alloys: IMI318/IMI685/Ti-6Al-4V / IMI550/IMI829/6-2-4-2S/TI

31-0407 METADEX Issue 8602

Superplastic Al—Cu—Li—Mg—Zr Alloys.

Lewis, R. E.; Pelton, A. R.; Wadsworth, J. Metall. Trans. A Dec. 1985; 16A, (12); 2319-2332; ISSN 0360-2133; in English

Superplastic properties have been developed in Al—Cu—Li—Mg—Zr alloys (Al—3Cu—2Li—1Mg—0.2Zr, Al—3Cu—2Li—1Mg—0.15Zr, Al—2.5Li—1.5Cu—1Mg—0.- 1Zr). The alloys have low Zr levels (<= 0.2 wt.%) and are experimental compositions that were originally designed as low-density, high-modulus, and high strength alloys for room temp., aerospace structural applications. The alloys have been manufactured both by an ingot metallurgy (IM) route and apowder metallurgy (PM) route involving rapid solidification processing. Afterconventional manufa cturing, the alloys are not superplastic but require furtherthermomechanical pr ocessing. The microstructural changes that occur during thisprocessing are des cribed. Superplastic properties have been evaluated as afunction of temp ., strain rate, and processing history. Prior tothermomechanical pr ocessing, the alloys have elevated-temp. ductilities of100-200%, strain r ate sensitivities of approx 0.25, and activation energiescorresponding to l attice diffusion. After thermomechanical processing, thealloys have ductil ities of 500-1000%, strain rate sensitivities of approx 0.4,and activation ener gies corresponding to grain boundary diffusion. In addition,room temp. propert ies have been measured in the solution-treated and peak aged(T6) condition for all the alloys and comparison is made with other commercial,non-Li containing, superplastic alloys. Particular emphasis is placed onproperties of inte rest in aerospace applications such as specific modulus,specific strength, a nd a buckling failure criterion. 52 ref.—AA

Category: 31. Controlled Terms: Alloying elements/Aluminum base alloys / Lithium/Mechanical properties/ Superplasticity/Zirconium. Alloys: Al-3Cu-2Li-1Mg-0.2Zr/Al-3Cu-2Li-1Mg-0.15Zr/Al-2.5Li-1.5Cu-1Mg-0.1-Zr/AL

8509-73987 Compendex

ALUMINUM AND TITANIUM COMPARED

Demmler, Albert W. Jr. Automotive Engineering, Warrendale, PA, USA. Automot Eng (Warrendale Pa) v 93 n 1 Jan 1985 p 51-59 Coden: AUEGB ISSN: 0097-711X In ENGLISH. Doc. Type: JOURNAL ARTICLE Treatment Des.: Applications THEORETICAL

The article discusses and compares the uses of aluminum and titanium in automotive and aerospace applications. Metallurgaical, corrosion resistance, formability, and cost consideration are mentioned

Card-A-Lert Codes: 541 (Aluminum and Alloys)/542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/655 (Spacecraft)/662 (Automobiles and Smaller Vehicles)/415 (Metals, Plastics, Wood and Other Structural Materials)/652 (Aircraft)/Controlled Terms: (*ALUMINUM AND ALLOYS — *Applications)/(TITANIUM AND ALLOYS — Applications)/(AUTOMOBILE MATERIALS — Aluminum)/(SPACECRAFT — Materials)/(AIRCRAFT MATERIALS — Titanium)/Uncontrolled Terms: FACE CENTERED CUBIC LATTICE/COMPOSITE OVERWRAPS/ SUPERPLASTIC FORMING/ARALL

8505-33397 Compendex
PRODUCTION PROGRESS WITH SUPERPLASTIC FORMING AND DIFFUSION BONDING AT BRITISH AEROSPACE
Anon. British Aerospace plc, London, Engl. Aircr Eng v 56 n 12 Dec 1984 p 5-6 Coden: AIENA ISSN: 0002-2667 In ENGLISH. Doc. Type: JOURNAL ARTICLE Treatment Des.: Applications

British Aerospace is pushing ahead with more production applications and further development of the superplastic forming and diffusion bonding (SPF/DB) titanium fabrication technique. This is in the wake of excellent results in series production of Airbus A310 wing components using this new method which gives major savings in weight and manufacturing cost. Confidence in the technical feasibility of applying the technique to manufacture of a wide range of complex structural aerospace components had led BAe to embark upon a multi-million pound SPF/DB introduciton program. By the end of 1984 the anticipated output of SPF and SPF/DB production components within BAe is expected to reach 5 tons/yr

Card-A-Lert Codes: 415 (Metals, Plastics, Wood and Other Structural Materials) / 652 (Aircraft)/541 (Aluminum and Alloys)/542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/535 (Rolling, Forging and Forming) Controlled Terms: (*AIRCRAFT MATERIALS — *Light Metals)/(TITANIUM AND ALLOYS — Forming)/Uncontrolled Terms: SUPERPLASTIC FORMING/ DIFFUSION BONDING

85N27727 NASA STAR Technical Report Issue 16
Characteristics and applications of superplastic materials for aeronautics
EIGENSCHAFTEN UND ANWENDUNGEN SUPERPLASTISCHER WERKSTOFFE FUER DIE LUFT-UND RAUMFAHRT
(AA)WINKLER, P. J. Messerschmitt-Boelkow-Blohm G.m.b.H., Ottobrunn (West Germany). (MT620643) Hauptabteilung Metalle. MBB-BB-565 84-O In its Res. and Develop. Tech. Sci. Repts. 1984 p 53-60 (SEE N85-27724 16-81) 840000 p. 8 refs 0 In: GM (German) Avail: Issuing Activity p.2873

Characteristics and applications of superplastic metallic materials are discussed. The technical requirements for superplastic materials and the composition of technically important superplastic alloys are given. Production techniques and economy of superplastic sheet metal forming are presented. Aeronautics and astronautics applications of aluminum alloys with average and high strength, and titanium alloys are described. Author (ESA)

Category code: 27 (nonmetallic materials). Controlled terms: *AEROSPACE INDUSTRY / *SUPERPLASTICITY *TECHNOLOGY UTILIZATION/ALUMINUM ALLOYS/METAL SHEETS / PRESSING (FORMING)/TITANIUM ALLOYS /

52-1347 METADEX Issue 8508
Low Weight, Low Cost SPFDB Parts.
Kellock, B. Mach. Prod. Eng. 6 Mar. 1985; 143, (3666); 18-20; ISSN 0024-919X; in English

Superplastic forming and diffusion bonding on Ti are considered. British Aerospace's pilot production facility at Warton is highlighted.—C.M.L.S.

Category: 52. Controlled Terms: Aircraft components/Diffusion welding/ Economics/Materials substitution/Metal working/Superplastic forming/ Titanium base alloys/Weight reduction. Alloys: Ti-6Al-4V/TI

85A27109 NASA IAA Conference Paper Issue 11
Superplastic behavior of aluminum-lithium alloys
(AA)WADSWORTH, J.; (AB)PALMER, I. G.; (AC)CROOKS, D. D.; (AD)LEWIS, R. E. (AD)(Lockheed Research Laboratories, Palo Alto, CA). IN: Aluminum-lithium alloys II; Proceedings of the Second International Aluminum-Lithium Conference, Monterey, CA, April 12-14, 1983 (A85-27101 11-26). Warrendale, PA, Metallurgical Society of AIME, 1984, p. 111-135. Research supported by Lockheed Independent Research Funds. 840000 p. 25 In: EN (English) p.0

Experimental results are presented on the superplastic behavior of two groups of alloys based on the Al-Li system. In the first group, additions of about 0.5 wt pct Zr were made to ingot alloys. In the second group, a thermomechanical processing procedure was carried out on two alloys, one produced by ingot metallurgy and the other by powder metallurgy. These alloys had similar compositions and contained only 0.15-0.2 wt pct Zr. The properties of the alloys are discussed, with emphasis placed on properties of interest in aerospace applications, such as specific modulus, specific strength, and a criterion for buckling failure. V.L.

Category code: 26 (metallic materials). Controlled terms: *ALUMINUM ALLOYS /*LITHIUM ALLOYS /*PLASTIC DEFORMATION /* SUPERPLASTICITY/MECHANICAL MEASUREMENT/SPECIMEN GEOMETRY/STRAIN RATE/TENSILE STRESS/ZIRCONIUM ALLOYS /

52-1212 METADEX Issue 8507
The Realization of SPF/DB as a Commercial Fabrication Process.
Weisert, E. D. Superplastic Forming; Los Angeles, California, USA; 22 Mar. 1984; American Society for Metals; Metals Park, Ohio 44073, USA; 1985; 8507-72-0310; 84-89; in English

Aerospace technology development initially involved only Ti, primarily the 6Al—4V alloy. More recently Al alloys have received considerable attention. Until the early part of the present decade, production applications of superplasticity in the aerosace industry have been sparse. Even now only an insignificant amount of the ultimate application potential of this sheet metal forming technology is being realized. Added stimulus to the application of superplasticity was provided by the invention involving diffusion bonding, known as SPF/DB. The progress of this combined Ti fabrication SPF/DB process toward commercial reality is considered. 5 ref.—AA

Category: 52. Controlled Terms: Aircraft components/Diffusion welding/Metal working/Sheet metal/Superplastic forming/Titanium base alloys. Alloys: Ti-6Al-4V/TI

52-1206 METADEX Issue 8507
Superplasticity in Aluminum Alloys.
Ghosh, A. K. Superplastic Forming; Los Angeles, California, USA; 22 Mar. 1984; American Society for Metals; Metals Park, Ohio 44073, USA; 1985; 8507-72-0310; 23-31; in English

A brief review of superplasticity in a number of Al alloys is presented. Emphasis is placed on an Al—Zn—Mg—Cu alloy (7475), which possesses high structural strength suitable for aerospace applications. Recent developments in thermomechanical processing have led to significant superplasticity in this material. Some new aspects of microstructural changes and mechanical behavior of this material are brought out. A short discussion on cavitation and its control is also presented. 20 ref.—AA

Category. 52. Controlled Terms: Aluminum base alloys/Grain size/Metal working/Microstructural effects/Strain rate/ Superplastic forming / Superplasticity. Alloys: 7475/Al-7Zn-2Cu-2.3Mg-0.12Cr/ AL

52-1205 METADEX Issue 8507
Superplasticity in Titanium Alloys.
Hamilton, C. H. Superplastic Forming; Los Angeles, California, USA; 22 Mar. 1984; American Society for Metals; Metals Park, Ohio 44073, USA; 1985; 8507-72-0310; 13-22; in English

Superplasticity of Ti has emerged as a viable manufacturing process over the past decade or so, and is now used to fabricate a number of sheet metal components for a range of aerospace systems. More than 200 parts are in production for a number of aircraft and spacecraft vehicles. An effort is made to provide a broad overview of factors understood to be important in the superplasticity of a range of Ti alloys. Since the bulk of the data are developed for Ti—6Al—4V, this alloy forms the primary basis for understanding of the factors important in superplasticity of Ti. However, available data for other alloys will guide the efforts to extrapolate the knowledge of Ti—Al—4V to a much broader spectrum of Ti alloys. A number of Ti alloys have been evaluated for superplastic behavior, and a summary of some of these alloys and their related superplastic properties is shown. 26 ref.—AA

Category: 52. Controlled Terms: Deep drawing/Grain size/Metal working / Microstructural effects/Strain rate/ Superplastic forming/Superplasticity / Titanium base alloys. Alloys: Ti-6Al-4V/Ti-6Al-2Sn-4Zr-2Mo / Ti-4.5Al-5Mo-1.5Cr/Ti-6Al-4V-2Ni/Ti-15V-3Cr-3Sn-3Al/Ti-15Mo/TI

52-1203 METADEX Issue 8507
Advances in Metals Processing.
Bales, T. T.; Royster, D. M. Advanced Materials Technology; Hampton, Virginia, USA; 16-17 Nov. 1982; National Aeronautics and Space Administration; Washington, DC 20546, USA; 1982; NASA CP-2251; 8507-72-0308; 201-218; in English

Research on metals processing is being conducted at the Langley Research Center to develop improved forming and joining methods with the potential of reducing the weight and cost of future aerospace structures. The approach followed is to assess the state of the art for fabricating a given structural system, define candidate methods for improving processing, evaluate the merits of each, fabricate and test subelement components, and then scale up the process to demonstrate validity. The development and the state of the art of weldbrazing, superplastic forming (SPF), superplastic forming and codiffusion bonding (SPF/DB) and superplastic forming and weldbrazing (SPF/WB) for Ti (Ti—6Al—4V) and the SPF of Al (3003, 7475) are reported. While the technology was developed for aerospace applications, potential uses are anticipated in the nonaerospace industries. 6 ref.—AA

Category: 52. Controlled Terms: Aircraft components/Aluminum base alloys / Brazing/Diffusion welding/Superplastic forming/Titanium base alloys. Alloys: 3003/7475/AL/Ti-6Al-4V/TI

52-1158 METADEX Issue 8506
Production Progress in Advanced Metalforming for Aerospace.
Metallurgia Jan. 1985; 52, (1); 16, 19, 20, 22, 23; ISSN 0141-8602; in English

The current position and recent advances in the production and development programme for Superplastic Forming and Diffusion Bonding at British Aerospace are outlined. The significant advantages and potential of the process within aerospace are highlighted, along with unique facilities employed and examples of parts produced for both civil and military aircraft, as well as missile and satellite field.—AA

Category: 52. Controlled Terms: Aircraft components/Corrosion resistance / Diffusion welding/Fabrication/Fatigue strength/Mechanical properties / Superplastic forming/Titanium base alloys/Weight reduction. Alloys: Ti-6Al-4V/TI

85A16087# NASA IAA Preprint Issue 05 **Technical and economical aspects in manufacturing aviation and space components by using superplastic forming (SPF) and/or superplastic forming/diffusion bonding (SPF/DB)**
(AA)BECK, W.; (AB)WINKLER, P.-J. (AA)(Messerschmitt-Boelkow-Blohm GmbH, Bremen, West Germany); (AB)(Messerschmitt-Boelkow-Blohm GmbH, Ottobrunn, West Germany). MBB-BB-572-84-OE Deutsche Gesellschaft fuer Metallkunde, Metallurgical Society of AIME, Akademiia Nauk SSSR, et al., International Conference on Titanium, 5th, Munich, West Germany, Sept. 10-14, 1984, Paper. 8 p. 840900 p. 8 refs 6 In: EN (English) p.0

The development of superplastic forming processes used in manufacturing aviation and space components, has been examined. Superplastic forming (SPF) and superplastic forming/diffusion bonding (SPF/DB) are carried out on Ti6Al4V annealed sheet metals. The effect of different process parameters, such as tooling and hot removal, on the quality of different components, is discussed. Process times can be reduced by hot charging, the use of a computer-aided process control system, and the use of ceramics. Related development activities for SPF and SPF/DB components have focused on the selection of suitable components, component redesign to suit SPF proce the definition of correct forming parameters and the final quality assurance. M.D.

Category code: 37 (mechanical engineering). Controlled terms: *AIRCRAFT STRUCTURES /*DESIGN TO COST / *DIFFUSION WELDING /* FORMING TECHNIQUES /*SPACECRAFT COMPONENTS /*SUPERPLASTICITY/AEROSPACE ENGINEERING/HOT WORKING / MISSILE BODIES/SANDWICH STRUCTURES/TITANIUM ALLOYS /

85A15611 NASA IAA Journal Article Issue 04
A primer on superplasticity in natural formulation
(AA)ARGYRIS, J. H.; (AB)ST. DOLTSINIS, J. (AB)(Stuttgart, Universitaet, Stuttgart, West Germany). Computer Methods in Applied Mechanics and Engineering (ISSN 0045-7825), vol. 46, Sept. 1984, p. 83-131. 840900 p. 49 refs 21 In: EN (English) p.0

A computer-oriented review of tasks arising in conjunction with superplastic deformation processes is presented. Material behavior under superplastic conditions is described, and an attempt is made to incorporate the influence of the rate of deformation and of growing grain size in the constitutive relation. In order to establish multiaxial relations, the natural approach to the mechanics of continua is reviewed and applied to the motion of incompressible viscous media arising under superplastic conditions. The appropriate multiaxial consitutive relations are then established in terms of the homogeneously defined natural stress and rate of deformation. Computational techniques for the analysis of slow viscous motion occurring in superplastic deformation processes are considered. Nonlinear solution methods for the velocity field are indicated and their convergence characteristics are demonstrated. A number of applications encompassing the bulging of circular membranes in titanium alloy and stainless steel are treated, and the forming of an actual space structure component in titanium is shown. C.D.

Category code: 26 (metallic materials). Controlled terms: *GRAIN SIZE /*PLASTIC DEFORMATION /*STRAIN RATE /* SUPERPLASTICITY /*TITANIUM ALLOYS/AEROSPACE INDUSTRY/CARTESIAN COORDINATES/ CONSTITUTIVE EQUATIONS/FINITE ELEMENT METHOD/GALERKIN METHOD /

52-0362 METADEX Issue 8503
Problems Involved in the Numerical Analysis of Plastic Working Processes for New Materials.
Rowe, G. W. Advanced Technology of Plasticity, Vol. 1; Tokyo, Japan; 1984; Japan Society for Technology of Plasticity; 5-2-5 Roppongi, Minato-ku, Tokyo 106, Japan; 1984; 8503-72-0109; 27-38; in English

The present state of knowledge in finite-element analysis of plastic working is discussed. The two major models of viscoelastic and elastic/plastic strain-hardening are reviewed. Examples are given of the application of FEM to forging, extrusion and compaction, under isothermal conditions and also when adiabatic heating is included. The difficulties of coupled thermomechanical analyses are considered. The properties of some superplastic and aerospace alloys (Mo-base and austenitic stainless steels) and the newer ceramics are discussed in the context of bulk forming. It is concluded that although these present formidable problems for numerical analysis, finite-element methods are in principle able to deal with such materials, and ways in which they may do so are suggested. 58 ref.—AA

Category: 52. Controlled Terms: Austenitic stainless steels/Backward extrusion / Compacting/Forging/Metal working/ Molybdenum base alloys/Numerical analysis/Plastic flow/Strain hardening

52-0254 METADEX Issue 8502
Properties and Applications of Superplastic Materials for Air and Space Travel.
Winkler, P.-J. Aluminium Apr. 1984; 6 (4); 261-266; ISSN 0002-6689; in German

Superplastic alloys Supral 220 (Al6Cu-.4Zr), AlZnMgCu and TiAl6V4 have mechanical properties particularly suitable for aerospace applications. Superplastic forming is very suitable for such uses; since the process is highly economic for small to medium series of complicated parts. The deformation behaviour of the alloys and the forming technology are outlined, together with examples of use. In the case of Ti structures the combination of superplastic forming and diffusion welding offers further design possibilities. 8 ref.—R.H.

Category: 52. Controlled Terms: Aerospace/Aluminum base alloys/Diffusion welding/End uses/Mechanical properties/Metal working/Superplastic forming/Titanium base alloys. Alloys: AlZnMgCu/2.4377/Al5Ca5Zn / Al6Cu0.4Zr/Supral 220/Supral 100/Supral 150/AL/TiAl6V4/3.7164/TI

8412-126898 Compendex
PROPERTIES AND APPLICATIONS OF SUPERPLASTIC MATERIALS FOR AEROSPACE
Winkler, Peter-J. Messerschmitt-Boelkow-Blohm GmbH, Central Lab, Munich, West Ger. Alum Engl v 60 n 4 Apr 1984 p 228-232 Coden: ALENE In ENGLISH Refs: 88 refs

Superplastic forming of aluminum and titanium sheet seems particularly well suited to aerospace applications, since the process is especially economical for small to medium-sized production runs of complicated articles of the type frequently encountered in the aerospace field. Nowadays, superplastic alloys such as Supral 220 (Al6Cu0. 4Zr), AlZnMgCu and TiAl6V4 are available, whose mechanical properties are suitable for use in the aerospace industries. Following a brief review of the deformation behavior of these alloys and the forming techniques involved, typical example applications are presented. With titanium structures, a combination of superplastic forming and diffusion welding offers additional design possibilities.

Card-A-Lert Codes: 415 (Metals, Plastics, Wood and Other Structural Materials) / 652 (Aircraft)/541 (Aluminum and Alloys)/542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/535 (Rolling, Forging and Forming) / Controlled Terms: (*AIRCRAFT MATERIALS — *Light Metals)/(LIGHT METALS — Superplasticity)/METAL FORMING /

84A49555 NASA IAA Conference Proceedings Issue 24
Superplastic forming — An economic sheet metal forming procedure
Superplastische Umformung — Ein wirtschaftliches Blechumformverfahren
(AA)WINKLER, P.-J. (AA)(Messerschmitt-Boelkow-Blohm GmbH, Zentrallabor, Ottobrunn, West Germany). IN: Sheet-metal forming: Fundamentals, technology, and materials (A84-49551 24-37). Oberursel, West Germany, Deutsche Gesselschaft fuer Metallkunde, 1983, p. 231-251. In German. 830000 p. 21 refs 7 In: GM (German) p.3510

The term 'superplasticity' refers to the ability of certain polycrystalline metallic materials to tolerate, when subjected to tensile stresses, extremely high elongation without fracture. During the last 10 or 15 years, attempts were made to utilize superplasticity for industrial applications. Superplasticity makes it possible to manufacture structural components with a complex geometry in one operation. In the present investigation, it is shown that, under certain conditions, the use of superplasticity can lead to a reduction of material costs and manufacturing costs. The characteristics of superplasticity are discussed, taking into account the dependence of the yield stress on strain hardening and the deformation rate. Attention is given to the determination of the m parameter, the characteristics of superplastic materials, the procedures used in superplastic sheet metal forming, economic considerations, and applications of superplastic forming in the aerospace industry. These applications are mainly based on the use of aluminum alloys of medium and high strength, and titanium alloys. G.R.

Category code: 26 (metallic materials). Controlled terms: *FORMING TECHNIQUES /*METAL SHEETS / *PLASTIC DEFORMATION /* POLYCRYSTALS /*SUPERPLASTICITY / AEROSPACE INDUSTRY/ALUMINUM ALLOYS/GRAIN SIZE/MECHANICAL PROPERTIES / TEMPERATURE DEPENDENCE/TITANIUM ALLOYS /

84A49551 NASA IAA Collected Work Issue 24
Sheet-metal forming — Fundamentals, technology, and materials
Blechumformung — Grundlagen, Technologie, Werkstoffe
(AA)LANGE, K. (AA)ED. (AA)(Stuttgart, Universitaet, Stuttgart, West Germany) Oberursel, West Germany, Deutsche Gesellschaft fuer Metallkunde, 1983, 414 p. In German. For individual items see A84-49552 to A84-49557. 830000 p. 414 In: GM (German) p.3554

The theoretical principles of deep drawing are considered along with the deformability of aluminum sheet material, metallurgical problems regarding the deformability of aluminum materials, the importance of superplastic forming as an economic sheet metal forming procedure, and the integrated manufacture of large-area sheet metal, taking into account operations from forming to contour finishing. Attention is also given to the significance of an application of strain analysis to sheet metal forming problems, the technological and economic significance of hydromechanical deep drawing, the employment of Computer Aided Design (CAD) in the development of an automobile model, the status of the technology of deep drawing and development trends, and possibilities and limitations in the case of a modification of material properties related to hot-roll-formed and cold-rolled-form steel. The forming of large sheet metal sections is also discussed, taking into account processing principles and examples related to the aerospace industry. G.R.

Category code: 37 (mechanical engineering). Controlled terms: *FORMING TECHNIQUES /*METAL SHEETS/ ALUMINUM ALLOYS/COLD WORKING/COMPUTER AIDED DESIGN/CONTOURS/DEEP DRAWING/ELASTIC BENDING/ELASTIC DEFORMATION/GRAIN SIZE/HOT WORKING/LARGE SPACE STRUCTURES/MICROSTRUCTURE/RESIDUAL STRESS / SHOT PEENING/STEELS/SUPERPLASTICITY/

46-0204 METADEX Issue 8412
Data Sheet. Transage 134 (Ti—2.5Al—12V—2Sn—6Zr) High-Strength Wrought Alloy.
Beta Titanium Alloys in the 1980's; Atlanta, Ga., U.S.A.; 8 Mar. 1983; The Metallurgical Society/AIME; 420 Commonwealth Dr., Warrendale, Pa. 15086, U.S.A.; 1984; 8412-72-0744; 485-487; in English

Transage 134 is a high-strength alloy useful for all wrought products, such as plate, bar, pipe and forgings. Metallurgically it is a martensitic alloy of exceptionally high hardenability. Because of exceptionally high resistance to corrosion and stress corrosion the alloy is useful for nonaerospace and aerospace applications. Transage 134 has exceptionally good fatigue properties. Due to the occurrence of strain-initiated, stress-induced transformation under conditions of high triaxial stress, it has exceptional resistance to fracture. A flow stress of < 70 MPa at 760 deg C and superplastic hot working characteristics qualify the alloy for net shape and near net shape forgings. Transage 134 has excellent welding and welding age hardening compatibility.—AA

Category: 46. Controlled Terms: Alloy development/Fracture toughness/Modulus of elasticity/Stress corrosion cracking/Tensile properties/Titanium base alloys. Alloys: Transage 134/Ti-2.5Al-12V-2Sn-6Zr/TI

84A35618 NASA IAA Journal Article Issue 16

Tensile properties of strongly textured Ti-6Al-4V after superplastic deformation

(AA)MCDARMAID, D. S.; (AB)BOWEN, A. W.; (AC)PARTRIDGE, P. G. (AC)(Royal Aircraft Establishment, Materials and Structures Dept., Farnborough, Hants., England). Materials Science and Engineering (ISSN 0025-5416), vol. 64, May 1984, p. 105-111. 840500 p. 7 refs 18 In: EN (English) p.2302

An attractive manufacturing process for the production of complex titanium alloy aerospace components is based on the utilization of superplastic forming (SPF). Titanium alloy components made by SPF and a diffusion bonding (DB) technique will be typically approximately 50 percent cheaper and 25 percent lighter than conventionally manufactured titanium alloy equivalents. The present investigation was undertaken to determine the effect of initial crystallographic texture and superplastic strain on the postforming room temperature tensile properties of the Ti-6Al-4V alloy. The deformation of strongly textured Ti-6Al-4V at 928 C under superplastic conditions was found to be anisotropic. G.R.

Category code: 26 (metallic materials). Controlled terms: *PLASTIC DEFORMATION /*SUPERPLASTICITY / *TENSILE PROPERTIES /*TITANIUM ALLOYS/ALUMINUM ALLOYS / BARS/CYCLIC LOADS/MICROSTRUCTURE/TENSILE STRENGTH/VANADIUM ALLOYS /

52-1719 METADEX Issue 8410

Superplastic Forming of Titanium and Aluminum Alloys for Aerospace Applications.

Berggreen, J.; Winkler, P. -J. Materials and Processes, Vol. 1; Monteux, Switzerland; 12-14 June 1984; Society for the Advancement of Material and Process Engineering; P.O. Box 2459, Azusa, Calif. 91722, U.S.A.; 1984; 8410-72-0649; Pp 11; in English

Several Al alloys (7475, Supral 100) and Ti alloys (Ti—6Al—4V) showing superplastic properties are important for aerospace application. This offers chances to fabricate complex shaped components in one piece and one step. After a short review to the metallurgical basis of superplasticity and diffusion bonding of these alloys the relevant processing parameters are presented. Parts from Ti and Al alloys are shown for space, helicopter and aircraft applications. These parts have been designed to SPF and fabricated. The process (e.g. oven or press) and the manufacturing process (e.g. design and material of the forming tool) have been optimized to cost and quality for each type of part. The state of production is reported.—AA

Category: 52. Controlled Terms: Aircraft components/Aluminum base alloys / Deep drawing/Die forming/Metal working/Sheet metal/Superplastic forming / Titaniumbase alloys. Alloys: 7475/Supral 100/AL/Ti-6Al-4V/TI

84A32686 NASA IAA Conference Paper Issue 14

Commercial applications of superplastic sheet forming

AA)SAWLE, R. (AA)(Superform Metals, Ltd., Worcester, England). IN: Superplastic forming of structural alloys; Proceedings of the Symposium, San Diego, CA, June 21-24, 1982 (A84-32676 14-26). Warrendale, PA, Metallurgical Society of AIME, 1982, p. 307-317. 820000 p. 11 In: EN (English) p.2017

Commercial exploitation of superplastic alloys is restricted to aluminum, titanium and zinc based systems. The economics of superplastic sheet forming ideally suit low to medium volume production. A typical product profile is discussed and examples of market areas using SUPRAL superplastic aluminum are given. The mechanical properties of the different alloy systems available largely determine their end use. Author

Category code: 26 (metallic materials). Controlled terms: *ALUMINUM ALLOYS /*FORMING TECHNIQUES / *METAL SHEETS /*METAL WORKING /*SUPERPLASTICITY /*TITANIUM ALLOYS/AEROSPACE INDUSTRY/MECHANICAL PROPERTIES/PRODUCT DEVELOPMENT/ZINC ALLOYS /

84A32685 NASA IAA Conference Paper Issue 14 **Aerospace applications of SPF and SPF/DB (superplastic forming with concurrent diffusion bonding)**

(AA)WILLIAMSON, J. R. (AA)(USAF, Materials Laboratory, Wright-Patterson AFB, OH). IN: Superplastic forming of structural alloys; Proceedings of the Symposium, San Diego, CA, June 21-24, 1982 (A84-32676 14-26). Warrendale, PA, Metallurgical Society of AIME, 1982, p. 291-306. 820000 p. 16 In: EN (English) p.1993

Development and production applications typical of efforts to establish the feasibility of superplastic forming (SPF) and superplastic forming with concurrent diffusion bonding (SPF/DB) of titanium structures are reviewed. The applications of SPF and SPF/DB in demonstration articles for manufacturing methods, design data components, structural validation components, flight test components, and production hardware are investigated. Typical cost savings of 30-50 percent and weight savings of 20-30 percent were realized in SPF/DB titanium structures such as a B-1 nacelle frame, a B-1 auxiliary power unit door, and a F-14A wing glove vane, as well as in a SPF YC-17 wing trailing edge flap skin. In a comparison of the conventional F-15 aft fuselage structure and the redesign of that structure for SPF and SPF/DB, it is shown that in the center-keel alone, 75 parts can be reduced to four parts and 1420 fasteners to 71 fasteners. The keel section itself was 77 percent less expensive. J.N.

Category code: 01 (aeronautics). Controlled terms: *AEROSPACE INDUSTRY /*AIRCRAFT PRODUCTION / *DIFFUSION WELDING /*FORMING TECHNIQUES /*SUPERPLASTICITY /*TITANIUM ALLOYS/ECONOMIC ANALYSIS/FLIGHT TESTS/PRODUCT DEVELOPMENT / STRUCTURAL DESIGN/TECHNOLOGICAL FORECASTING /

84A32676 NASA IAA Meeting Paper Issue 14
Superplastic forming of structural alloys; Proceedings of the Symposium, San Diego, CA, June 21-24, 1982
(AA)PATON, N. E.; (AB)HAMILTON, C. H. (AA)ED.; (AB)ED. (AA)(Rockwell International Corp., Pittsburgh, PA); (AB)(Rockwell International Science Center, Thousand Oaks, CA). Symposium sponsored by the Metallurgical Society of AIME and American Society for Metals. Warrendale, PA, Metallurgical Society of AIME, 1982, 424 p. 820000 p. 424 In: EN (English) Members, $24.; nonmembers, $36 p.2016

Research on superplasticity as a low-cost production method for the manufacture of complex components is presented in terms of basic mechanisms, superplastic materials, superplastic forming processes and applications, and cavitation in superplastic alloys. Specific areas of study include the superplastic behavior of metals; superplasticity in titanium-base alloys, high-strength aluminum alloys, and nickel-base alloys; and superplastic forming of sheet metal. Attention is also given to aerospace applications of SPF and SPF/DB and commercial applications of superplastic sheet forming. For individual items see A84-32677 to A84-32687 J.N.

Category code: 26 (metallic materials). Controlled terms: *CONFERENCES /*FORMING TECHNIQUES /*HEAT RESISTANT ALLOYS /* HIGH STRENGTH ALLOYS /*METAL WORKING /*SUPERPLASTICITY/AEROSPACE INDUSTRY/AIRCRAFT STRUCTURES/ALUMINUM ALLOYS / DESTRUCTIVE TESTS/GRAIN BOUNDARIES/ MECHANICAL PROPERTIES/METAL SHEETS / NICKEL ALLOYS/PLASTIC FLOW/PRODUCT DEVELOPMENT/STRAIN RATE/TITANIUM ALLOYS /

61-0639 METADEX Issue 8409
Stiffened Panel.
Mansbridge, M. H.; Norton, J. British Aerospace. 16 May 1984; 8 Nov. 1983; Patent no: GB2129340A; UK

A method of making a stiffened panel includes subjecting two metal sheets, at least one of which is capable of superplastic deformation and diffusion bonding, and which are positioned face to face, to a bonding and deforming process during which the sheets are joined to one another at a series of spaced joint lines across their faces. The joint lines are interrupted by non-joined regions along their lengths but are otherwise substantially continuous. Parts of at least one of the sheets between the joint lines and the non-joined regions thereof are superplastically deformed in a mould to form a series of cavities between the two sheets. Portions of at least one of the sheets on respective sides of each of the joint lines and the non-joined regions are moved to form sidewalls of two neighbouring cavities, these sidewalls being urged to lie adjacent to one another over substantial parts of their areas, so that they become diffusion bonded, one to another, to form a common sidewall of neighbouring cavities. The non-joined regions of the joint lines each form a generally circular or part-circular aperture in each sidewall of a diameter similar tothe length of the non-joined region.

Category: 61. Controlled Terms: Diffusion welding/Fabrication/Metal working / Panels/Sheet metal/Stiffening/ Superplastic forming

84A30750 NASA IAA Journal Article Issue 13
Superplastic behavior of a powder-source aluminum-lithium based alloy
(AA)WADSWORTH, J.; (AB)PELTON, A. R. (AA)(Lockheed Research Laboratories, Palo Alto, CA); (AB)(U.S. Department of Energy, Ames Laboratory, Ames, IA). Scripta Metallurgica (ISSN 0036-9748), vol. 18, April 1984, p. 387-392. Research supported by the Lockheed Independent Research and Development Fund. 840400 p. 6 refs 15 In: EN (English) p.1861

The development of Al-Li alloys is currently considered with great interest. This situation exists because the addition of lithium to aluminum provides benefits of improved elastic modulus and decreased density. For these reasons, alloys based on Al-Li could offer significant weight savings in aircraft and other aerospace applications in comparison to designs employing conventional aluminum alloys. Research and development programs have mainly been concerned with allo y producibility and ambient temperature mechanical properties. Starke and Sanders (1983) and Wadsworth et al. (1983) have now, however, studied the formability of the alloys at elevated temperatures. Mehrabian (1983) has considered powder metallurgy (PM) methods and the use of Rapid Solidification Processing (RSP). In the present study, the high-temperature deformation characteristics have been determined for Al-Li based alloys prepared by RSP. A description of the deformation behavior is given both for material in the consolidated condition and for material which has been subsequently thermomechanically processed. G.R.

Category code: 26 (metallic materials). Controlled terms: *ALUMINUM ALLOYS /*HIGH TEMPERATURE TESTS /*LITHIUM ALLOYS /*PLASTIC DEFORMATION /*POWDER METALLURGY /*SUPERPLASTICITY/GRAIN BOUNDARIES/SOLIDIFICATION/STRAIN RATE / STRESS-STRAIN RELATIONSHIPS/STRUCTURAL WEIGHT / THERMOMECHANICAL TREATMENT /

8404-38255 Compendex
SUPERPLASTIC FORMING AND DIFFUSION BONDING
Anon. Aircr Eng v 55 n 9 Sep 1983 p 22 Coden: AIENA ISSN: 0002-2667 In ENGLISH

A purpose-built superplastic forming and diffusion bonding manufacturing facility has been commissioned at the Hatfield Division of British Aerospace Dynamics Group. Components are being made by the superplastic forming and diffusion bonding technique from titanium, and the method is also being developed for application to the manufacture of items from special aluminum and stainless steel alloy sheet materials. The technique offers several major advantages of which simplicity and economy of manufacture are among the more important. Complex components can be made in one piece, in one operation, leading to a reduction in the number of parts required and also the elimination of subsequent assembly operations. Excellent dimensional accuracy and repeatability of form is achieved

Card-A-Lert Codes: 542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/535 (Rolling, Forging and Forming)/415 (Metals, Plastics, Wood and Other Structural Materials)/652 (Aircraft)/Controlled Terms: (*TITANIUM AND ALLOYS — *Forming)/(AIRCRAFT MATERIALS — Superplasticity) /

54-0488 METADEX Issue 8408
Emerging Trends in Aerospace Materials and Processes.
Chandler, H. E. Met. Prog. Apr. 1984; 125, (5); 21-29; ISSN 0026-0665; in English

Current and expected future trends in aerospace materials and processes are discussed. Wrought Al is forecast to hold its' position, but with increasing competition from composites and Al—Li alloys. The future for Al P/M applications looks hopeful, though commercial usage is not yet widespread. The use of Ti in aircraft structures is expected to enjoy a steady, if unspectacular, growth.—G.P.K.

Category: 54. Controlled Terms: Aircraft components/Alloying elements / Aluminum base alloys/Lithium/Materials substitution/Metal working/Nickel base alloys/Powder metallurgy parts/Powder technology/Superalloys / Superplastic forming. Alloys: 7075/2024/7090/7091/AL/HP310/SAHS / Udimet 720/Udimet 700/Incoloy 903/Incoloy 907/Incoloy 909/Rene 95 / Hastelloy X/PWA-1056 / MERL-76/NI/SP/Ti-10V-2Fe-3Al / Ti-15V-3Al-3Cr-3Sn/TI

71-0255 METADEX Issue 8406
Future Uses of Aluminium Alloys.
Woodward, A. R. The Metallurgy of Light Alloys.; Loughborough University, England; 24-26 Mar. 1983; The Institution of Metallurgists; P.O. Box 471, 1 Carlton House Terrace, London SW1Y 5DB, England; 1983; 8406-72-0435; 1-8; in English

The versatility of aluminium and its alloys is stressed and explained in terms of their frequent use as a substitute material, excepting the case of airframe construction where they have been predominant. Developments in the systems Al—Zn—Mg, Al—Mg—Si and Al—Mg alloys in the structural, architectural and particularly the packaging field for uses such as beverage cans are described. The inability to adapt Al—Cu—Mg—Si or Al—Zr—Mg—Cu alloys used for airframes and aerospace purposes to other uses is explained by their relative lack of weldability. Some final points are made about formability of sheet alloys, the availability of alloys suitable for superplastic forming and a brief mention is given of developments in cast products.—AA

Category: 71. Controlled Terms: Aluminum/Aluminum base alloys/Development / End uses

71-0177 METADEX Issue 8404
Titanium.
Roberts, W. T. Endeavour 1983; 7, (4); 189-193; ISSN 0160-9327; in English

Types and locations of Ti ores are reviewed and present and potential routes for their reduction to powders and granules and the production of ingots, forgings, flats, bars, rods, wires, tubes, compacts, and castings are flow-charted and discussed. Large-scale production of the sponge or granules from TiCl sub 4 is currently carried out only in the USSR, the USA, Japan, and the UK, with annual sponge capacities of 40 000, 25 000, 15 000, and 5000 tonnes, resp. The batch Kroll and Hunter processes for reducing the tetrachloride are described and outlines given of vacuum arc remelting for the production of sound Ti or Ti-alloy ingots and press-forging these to a range of semi-fabricated forms, the proportionate outputs of which are specified. Characteristics of the alpha , beta , and alpha + beta alloys are given with ref. to microstructural/property relationships of these and of the unalloyed metal. The differing requirements for aerospace and general engineering applications are outlined with mention of isothermal forging, superplastic forming, and diffusion bonding. The mechanism of the excellent corrosion resistance in a wide range of environments is also considered, together with their exploitation in the chemical, paper, pulp, textile, powder, metallurgical, and marine-engineering industries and in surgical, desalination, etc., applications. Illustrations include the creep behaviours of various Ti alloys and use of these in aero-engine and helicopter assemblies.—J.R.

Category: 71. Controlled Terms: End uses/Extraction/Productivity/Refining / Titanium/Titanium base alloys/Titanium ores

52-0593 METADEX Issue 8404
Superplastic Forming Makes Molded Parts From Sheet Metal Alloys.
Hamilton, C. H. Res. Dev. Dec. 1983; 25, (12); 72-76; ISSN 0160-4074; in English

Superplastic alloys exhibit exceptionally high tensile ductility. They can be stretched to elongations from 300 to 3000% at a proper temp. and within a controlled strain rate range. There is a substantial reduction in the flow stress of the material, 0.5 to 0.05 of that of a conventional alloy. Fundamental requirements for superplasticity are a fine, stable grain size, a deformation temp. that is in excess of half of the absolute melting point and a carefully controlled rate of straining. Two-phase Ti alloys are suitable when conventionally processed, others, such as Al alloy 7475, must be specially processed. Other superplastic alloys are: Ni P/M IN-100, Al Supral 100 and 08050, Zn—22Al and Fe IN744. SPF can be combined with diffusion bonding to produce even more complex parts in one operation. Applications are in the aerospace industry where major weight and cost savings have been realized.—MAY

Category: 52. Controlled Terms: Aircraft components/Ferrous alloys/Metal working/Nonferrous alloys/Sheet metal/ Superplastic forming. Alloys: Supral 100/7475/AL/Ti-6Al-4V/TI/Zn-22Al/ZN/IN100/NI/IN744/FE

8401-8512 Compendex
FORM-BONDING TITANIUM IN ONE SHOT
Vaccari, John A. American Machinist, New York, NY, USA. Am Mach v 127 n 10 Oct 1983 p 91-94 Coden: AMMAA ISSN: 0002-9858 In ENGLISH

With the Ti-6Al-4V alloy, superplastic-forming (SPF) temperatures are amenable to the alloy's diffusion-bonding (DB) temperatures, permitting simultaneous superplastic forming and diffusion bonding (SPF/DB) of multisheet packs into truss-core structures. Although SPF characteristics alone can drastically reduce the number of individual parts required for various components, SPF/DB couples this advantage with the structural benefits and cost-effectiveness of such lightweight sandwich structures. Superplastic forming and simultaneous superplastic forming and diffusion bonding of Ti-6Al-4V are catching on in the aerospace industry

Card-A-Lert Codes: 542 (Beryllium, Magnesium, Titanium and Other Light Metals and Alloys)/541 (Aluminum and Alloys)/543 (Chromium, Manganese, Molybdenum, Tantalum, Tungsten, Vanadium and Alloys)/535 (Rolling, Forging and Forming)/

Controlled Terms: (*TITANIUM ALUMINUM VANADIUM ALLOYS — *Bonding)/(SHEET AND STRIP METAL — Forming)/Uncontrolled Terms: SUPERPLASTIC FORMING / DIFFUSION BONDING

52-0290 METADEX Issue 8402
Superplastics Set for Take-Off.
New Sci. 8 Sept. 1983; 99, (1374); 687; ISSN 0028-6664; in English

The superplasticity of Ti at approx. 950 deg C will be exploited at the Hatfield works of the British Aerospace Dynamics Group to effect diffusion-bonding for the production of complex shapes without milling or welding. Moulds charged with Ti pieces will pass through four main robot-served stages, comprising loading and unloading before and after heating, presentation to a 150 tonne press, and deposition in a cooling area, the only subsequent treatment being an acid dip to remove minor contamination and excess metal. In full operation the plant will work automatically, including overnight and weekend periods. Despite the greater cost of Ti the shaped components will be 45% cheaper than Al counterparts as well as being up to 25% lighter, and can replace steel with weight and cost savings of up to 45 and 65%, resp.—AA

Category: 52. Controlled Terms: Automation/Diffusion welding/Industrial robots/Materials substitution/Metal working/Superplastic forming / Titanium

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14. Abstract

(transformed)
Superplasticity has been transormed from a metallurgical curiosity to an important production process, particularly for low-to-medium production runs of components for the aerospace industry.

The whole spectrum of superplasticity was originally covered in Lecture Series 154 in the autumn of 1987, but such are the rapid advances in this technology, that the series is re-presented, with the same speakers updating their lectures to give those attending the latest information on this most relevant technology and its impact upon the manufacture methods employed for components for aerospace applications. Keywords:)

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